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RESEARCH MEMORANDUM

NOTES ON HEAT-RESISTANT MATERIALS IN BRITAIN FROM TECHNICAL MISSION OCTOBER 13 TO NOVEMBER 30, 1950

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NATIONAL ADVISORY COMMITTEE FOR AERONAUTICS

RESEARCH MEMORANDUM

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By J. W. Freeman and Howard C. Cross

SUMMARY

This report summarizes information on heat-resistant materials obtained from interviews with alloy producers, jet-engine manufacturers, and research laboratories in Britain during October and November 1950. Attention was directed to the relationship between properties of materials and service requirements for gas turbines; to criterions used for judging and inspecting materials; and to research developments of both a practical and fundamental nature in the field. Some information is also included regarding ceramics, ceramals, and materials for high-temperature applications in steam-power plants.

INTRODUCTION

A technical mission to Britain was made from October 13 to November 30, 1950, to obtain information on the status of research and development in heat-resistant materials. This mission was under the sponsorship of the Directorate of Research and Development, Headquarters, U. S. Air Force; the Office of Naval Research, Department of the Navy; and the National Advisory Committee for Aeronautics. This report summarizes the information obtained by the authors from interviews with alloy producers, jet-engine manufacturers, gas-turbine manufacturers, and research laboratories. Attention was directed to the relationship between properties of materials and service requirements for gas turbines; to the criterions used for judging and inspecting materials; and to research developments of both a practical and fundamental nature in the field. The report is primarily of a metallurgical nature, although some ceramic and ceramal information is included. Some information is also included regarding materials for high-temperature applications in steam-power plants.

In general, the information does not include that published by the British Iron and Steel Institute as a Symposium on High-Temperature

Steels and Alloys for Gas Turbines held during February 1951. Many details which would normally have been included in this report can be found in the comprehensive papers of this Symposium.

The establishments interviewed were:

Department of Aeronautical and Engineering Research

Armstrong Siddeley's, Ltd.

The Bristol Aeroplane Company, Ltd.

Bristol University

British Non-Ferrous Metals Research Association

The Brown-Firth Research Laboratories

Cambridge University (Department of Metallurgy)

Celvilles, Ltd.

De Havilland Aircraft Company, Ltd.

Fulmer Research Institute

Wm. Jessop and Sons, Ltd.

King's College, University of Durham

Metal-Gas Company, Ltd.

Metropolitan-Vickers Electrical Company, Ltd.

Ministry of Supply

The Mond Nickel Company, Ltd.

The Mond Nickel Company, Ltd., Development and Research Department Laboratory

National Gas Turbine Establishment

National Physical Laboratory

C. A. Parsons and Company, Ltd.

Pametrada Research Station of the Parsons and Marine Engineering Turbine Research and Development Association

Rolls-Royce, Ltd.

Royal Aircraft Establishment

The United Steel Companies, Ltd.

GAS-TURBINE BLADES

Nimonic 90 is being used in production jet engines for either higher temperature or higher stress service than Nimonic 80A will stand. Present practice is to use Nimonic 80A for all blading except for the more severe service conditions.

In multistage turbines for turboprop engines, alloys considered to be lower down the scale in strength are used for the lower temperature rows of blading. G18B and Nimonic 80 are used for this purpose.

The practice for power gas turbines generally follows that of the aircraft field except that Nimonic 90 is not used, because the temperature and stresses are not high enough in present designs to require its use, and because there is some uncertainty whether its long-time properties are superior to those of Nimonic 80A and other alloys. In at least one case, the ferritic Mo-V steel is used for the lower temperature blades.

The specified composition in Nimonic 90 is

Carbon	0.1 max.
Silicon	1.5 max.
Manganese	1.0 max.
Chromium	18-21
Cobalt	15-21
Titanium	1.8-2.7 (aim 2.4)
Aluminum	0.8-1.8 (aim 1.2)
Iron	5.0 max. (usual 0.3)
Nickel	Balance
Sulfur content of chromium	0.05 max.

Rolled bar stock is furnished to the users from 150-pound ingots. The recommended heat treatment is:

(a) Solution-treated 8 to 12 hours at 1975° F, air-cooled

(b) Aged 12 to 16 hours at 1290° F

A small ingot is poured from each heat, rolled to bar stock, heat-treated, and creep-rupture tested before the remainder of the heat is processed. At present, the internal control test for a heat is that third-stage creep must not occur before 75 hours nor rupture before 100 hours, with a minimum creep rate of not more than 0.01 percent per hour at 1382° F under 42,500 psi. This test is known to insure claimed properties out to 1000 hours and is advocated as insuring claimed properties at higher temperatures.

A great deal of experience must be accumulated on an alloy relating short-time creep behavior to 1000-hour behavior before one test can be relied upon for quality checking. Tests of about 100 hours are being used by the Mond Nickel Company to establish that their alloys have at least satisfactory minimum strength for 0.1-percent deformation in 1000 hours. They think that trace elements, principally present in the rather impure cobalt available for use in Nimonic 90, are responsible for the occasional heats that are deficient in creep properties. If electrolytic cobalt of much higher purity were available, Nimonic 90 could be made with more consistent properties.

The occasional blade failures by fatigue which have been examined by The Mond Nickel Company have been found to be accompanied by third-stage creep. They feel that gas-turbine blades fail by fatigue only after the onset of third-stage creep, unless the design is at fault. Because both fatigue failure and excessive deformation limit the usefulness of blades, and both are largely controlled by the onset of third-stage creep, this property of an alloy is of prime importance to the designer.

Specified properties for the manufacturer's check test are the same, except that the minimum time for start of third-stage creep is 50 hours, and the minimum time for rupture is 75 hours.

Industrial acceptance tests are generally made for the purpose of assuring the user that the material being used, particularly forged blades, will meet the probable maximum operating conditions in an expected total life of 150 hours in an engine. Some users check the properties of the bar stock, while others run checks on specimens cut from rough blade forgings.

There is considerable pressure from industry to increase the acceptance test requirements for Nimonic 90. One suggested requirement is that third-stage creep not occur in less than 100 hours at 1500° F under 29,000 psi, and it is probable that this requirement can be met in all except a few heats. Another suggested requirement is a minimum rupture time of 100 hours under 47,000 psi at 1380° F, with the indications being that this could easily be met.

The users' acceptance tests vary considerably among engine manufacturers. This is the result of different philosophies as to engine design and of different opinions as to what are the controlling factors in engine operation which determine the life of the components. For instance, Rolls-Royce feel that the controlling factors are the peak temperatures and stresses which occur under take-off and emergency power conditions, such as in combat. Their acceptance test is for 30 hours at 1600° F, the total time they estimate the turbine will be called upon to operate under peak conditions during the total estimated times of peak performance of the engine. On the other hand, while production blades are checked by De Havilland with a short-time rupture test, they choose their blade materials and use them under conditions which will give about a 90° F margin below the 1000-hour rupture strength. Because of their desire to build reliable long-life engines, they feel that the 90° F margin based on the 1000-hour rupture will adequately compensate for variability in both materials and engine operating conditions.

Nimonic 90 was generally rated as being about 80° to 90° F superior to Nimonic 80A in engine service. Generally, Nimonic 90 was reported to have higher fatigue strengths than Nimonic 80A, but this opinion was not unanimous, possibly because of different conditions of fatigue testing, such as temperatures, stresses, and types of testing machines.

By as carefully working out and controlling production and processing techniques for Nimonic 90 as had been done previously for Nimonic 80A, an alloy of consistent properties has been produced. The occasional lot that does not come up to the specification requirement is eliminated as a result of the producers' tests before final processing. Fabricating characteristics and general handling during processing into turbine blades indicate the alloy is no more difficult to handle than Nimonic 80A. Nimonic 90 blades are forged using only one blow per reheat. In cases where improper machining and finishing operations resulted in a cold-worked surface on the finished blade, blade failures have occurred in service.

As an example of one engine producer's method of handling Nimonic 90, blades are forged oversize (1/8 to 1/16 in.) by De Havilland. The standard solution treatment is then carried out. The rough machining operation of milling the convex side and slotting the concave side is carried out, leaving the blade about 0.015 inch oversize. The blades are then reheated to the solution-treatment temperature for 15 minutes and water-quenched. Then, the final 0.015 inch is machined off, with care taken to minimize cold-work. By final electromachining and then aging, any residual cold-work is reduced below a harmful level.

A much coarser grain size for blades of Nimonic alloys is used in Britain than is acceptable in the United States. Nimonic 90 is usually supplied with an A.S.T.M. grain size of zero, whereas American practice requires A.S.T.M. 6 to 8. The coarse-grain-size material shows no reduction in fatigue strength, but mixed grain size is indicative of a condition that gives trouble under fatigue stressing. The levels of creep-rupture strength required for Nimonic 90 cannot be attained with fine-grained material. Raising the solution-treatment temperature from 1980° to 2100° F so as to increase further the creep-rupture strength is being considered by The Mond Nickel Company.

Mixed grain sizes in Nimonic alloys are indicative of poor properties, not because the grain size is mixed, but as a result of the conditions producing the mixed grain size. Patches of mixed grain size are particularly poor in fatigue.

Poor forging practice is one cause of mixed grain sizes. Conditions setting up strain gradients are causes, particularly forging at too low temperatures and not sufficiently heavy reduction. Improper solution treatment and incomplete solution of precipitates in heating for forging set up obstructions to grain growth causing mixed grain size.

Nimonic 80A is being used in production engines for all blading in the Rolls-Royce Nene, the Armstrong-Siddeley Mamba and Python, and the Bristol Proteus engines. These stages operate at lower temperatures than the first-stage rows bladed with Nimonic 90.

As far as is known, no jet aircraft engine is, at the present, using any other blading alloy of strength characteristics lower than those of Nimonic 80. All of the factors such as production, processing, heat-treating, and finishing mentioned during the discussion of Nimonic 90 apply equally to Nimonic 80A and 80.

For operation at lower temperature levels, both in the latter stages of the Bristol turboprop Proteus engine or in the several large power turbines being constructed or run in tests, alloys with lower strength levels such as the austenitic alloys G18B, R.ex 326F, FCB(T) (18Cr, 8Ni, 1.25Cb), and R20(19Cr, 14Ni, 1.7Cb) and the ferritic Mo-V steel are used. The available properties for these alloys are included in the appendix. The Mo-V steel is usable up to about 1050° F, but above that temperature its oxidation resistance is not adequate, and the austenitic materials are required.

Some work is in progress on development of blade alloys with properties superior to those of Nimonic 90. Alloy G32, a cobalt-base alloy with about 0.3 percent carbon, shows some promise in the temperature range above 1500° F and shows about the same strength as Nimonic 90. An investment-casting modification containing 0.8 percent carbon, identified as G34, is being evaluated. The properties for G32 alloy are included in the appendix. No data are available for G34.

[redacted]

No cast alloys are in use for rotating turbine blading. Considerable experimental work is in progress on cast alloys in an effort to produce consistently good-quality alloy and to reduce the scatter in properties. One turbine producer felt quite strongly that an aggressive program to develop investment casting of blades should be started. Emergency requirements of Nimonic 90 and 80A alloys might be difficult to meet, and forging and machining requirements would also be overtaxed. The availability of easily expandable investment-casting techniques would be very important then. There was some disagreement as to the relative costs of cast or machined blades, but it was indicated that cast blades would deserve more consideration on the basis of a large production, because then the cost might be more competitive. The use of cast blades affords an easy way to make changes in blade design.

The Nimonic alloys containing titanium and aluminum are very difficult to cast. No cobalt-free alloys that will give good properties in the as-cast condition have been found.

Conversations indicated that other alloys are being studied, but no data could be obtained. These improved alloys of the future may be the unforgeable alloys of the present. Some success may be achieved in this direction by extrusion rather than by other methods of processing. The indications are, however, that alloys with promising strength at temperatures above those currently used are prone to be brittle at room temperature.

One item of interest to producers of forged turbine blades is the Brown-Firth research on the problem of abnormal grain growth in solution-heat-treated wrought austenitic steels. This problem has been studied intensively by Brown-Firth. They have concluded that it is associated with hot-working at temperatures in the order of about 1800° to 1840° F, with the temperature of working being much more important than the percentage reduction. These temperatures are just under the lowest temperatures for simultaneous recrystallization during hot-working. The closer the hot-working temperature approaches this recrystallization temperature, the larger are the grain sizes subsequently produced on solution heat treatment. If the recrystallization temperature is slightly exceeded during the hot-working, solution treatment produces very fine grains.

These principles have been developed in studies of their austenitic rotor disk forgings which are of such large mass that no large temperature gradients are obtained in the material during hot-working. They do not understand at this time how to apply these principles so as to prevent the formation of large grains during solution treatment in parts of such mass that, during normal hot-working operations, an appreciable temperature gradient results.

TURBINE STATOR BLADES

Information on alloys for turbine stator blading was obtained only on the jet aircraft engines and not for turboprop engines or the various power gas turbines. Generally, the alloys being used were not so satisfactory as desired. Some wrought H.R. Crown Max (0.35C, 23Cr, 12Ni, 3W) was being used, but its thermal-shock resistance is marginal and not so good as that of H.S. 21. H.S. 21 shows about a 3 to 1 superiority in thermal-shock resistance and is the best nozzle vane material for temperatures of 1650° F (900° C) and above. Stator blades of H.R. Crown Max require about 20 percent replacement after 300 hours of service.

Several jet aircraft engine producers use Nimonic 80 for stator blades. One producer uses investment castings and another, wrought alloy. The cast Nimonic 80 does show some trouble from thermal cracking and some trials with ceramic coatings are being made in an effort to eliminate this trouble. Wrought Nimonic 80 is used by De Havilland not so much from preference (Nimonic 75, for instance, being considered as good) as to use up lots of the alloys that are not sufficiently strong in creep resistance to pass as Nimonic 80A. The use of hollow cast turbine stator blades of modified H.R. Crown Max is being investigated by De Havilland. This alloy has shown the best casting characteristics among all of the alloys they have tried. The hollow blade is expected to have better temperature uniformity than a solid blade, with consequent improvement in performance.

The general impression gained is that better stator blade performance is needed, and the use of ceramic coatings and hollow blades is directed toward that end. Comments were made that cast H.S. 21 is significantly better than the presently used materials and that their difficulties would be eliminated if this cobalt-base alloy were permitted to be used.

At Pametrada it is expected that ceramics can be used for fixed blades and they have been experimenting with zircon-base ceramic nozzle vane blades. Compositional changes are being made to improve the thermal-shock resistance, which is apparently not adequate.

COOLED BLADES

Blade cooling for rotating blades of turbines is as yet a matter of design. For this reason, and because discussions were largely with metallurgists, little information was obtained. Air cooling certainly has been carefully reviewed. The use of liquid cooling is probably receiving considerable attention, at least for land turbines. No indication of materials for cooled blades was obtained beyond the fact that experimental blades were being made from Nimonic alloys by drilling and various attempts were being made to cast hollow blades.

Experimental hollow stator blade castings were exhibited by De Havilland. There were other indications that cooling of stator blades was being considered. Sweat cooling is receiving attention on an experimental basis. The only direct contact with a material with possible application in this field was the porous 18-12+Cb material which Wm. Jessop and Sons are developing. They claim to have solved the problem of controlled pore size and distribution for this alloy with near theoretical physical properties using powder-metallurgy techniques. Even for this alloy, practical application is being held up, pending development of suitable welding or other joining methods. No indication of a practical solution to pore-clogging problems in connection with sweat cooling was observed.

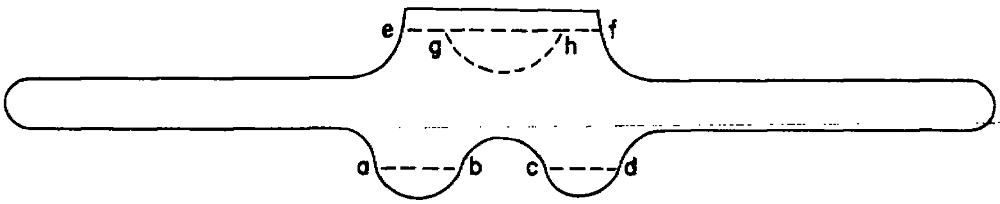
AIRCRAFT GAS-TURBINE DISKS

No one is requesting a better austenitic material than G18B with regard to creep resistance, but most want a material with better room-temperature proof stress, lower cost and thermal expansion, and easier to manipulate in manufacture. The trend is away from austenitic disks and into ferritic disks. They are easier to machine, less susceptible to thermal shock, and cost less. The successful use of ferritic steels requires adequate air cooling. In one case, consideration was being given to age-hardenable alloys of the Discaloy type.

The alloys are described in tables I through IV. Typical high-temperature properties are summarized in figure 1. The 12Cr steels, R.ex 448 and H46, are relatively new developments which maintain high strength up to 1200° F. R.ex 448 has pertinent high-temperature properties up to 1200° F, slightly better than those of 18-12+Cb steel. The properties reported for H46 are somewhat lower, but an "improved" H46 was indicated to have properties of the same order as 18-12+Cb. It was indicated that large-scale commercial production of the 12Cr steels had been achieved with uniform properties and good quality. The analysis and heat treatment of R.ex 448 were not disclosed nor was the change in H46 to achieve similar properties.

Inspection of disks for aircraft turbines is mainly dependent on macroetching and surface inspection. Tensile specimens for checking physical properties at room temperature are taken from excess metal provided at the hub of rough forgings for this purpose. Ultrasonic inspection is not a specification requirement, although disk producers generally use it for developing and insuring sound disks.

In one case, disks are rough forged to the following type shape:



Slice a-b-c-d is removed by the disk producer to check tensile properties. Slice e-f is removed by the engine manufacturer to check tensile properties. The contour g-h is then machined into the disk. The whole disk is then macroetched. The removal of metal to form cavity g-h exposes the most vulnerable part of the forging for poor quality to macroetching. All tensile specimens must have, for 3Cr-Mo-W-V steel, a minimum yield strength for 0.1-percent offset of 90,000 psi, with 15 percent elongation at room temperature. Specimens are occasionally taken from the rims of disks for checking creep resistance, the requirement being less than 0.5 percent creep in 300 hours at 1112° F under 24,640 psi.

Only two companies were found to be using overspeed testing of disks. At Bristol it is used for the dual purpose of inspecting of G18B disks and raising yield strength. Armstrong Siddeley also inspect Mamba and Python G18B disks by overspeeding. Overspeed inspection of ferritic disks will be continued by this company until experience indicates it is unnecessary.

Some difficulty has been encountered in development engines as a result of fatigue failures at blade-root serrations. This was caused by a poor fit between disk and blade serrations, causing stress concentrations at one point on the lands of the blade serrations.

The experience quoted for disks is difficult to generalize. The following is a brief summary for the various alloys.

Mo-V steel.- This steel is generally considered to have good creep resistance up to 1100° F. Poor oxidation resistance, however, limits its use to 1025° to 1050° F. The greatest criticism of the alloy was low hardenability, resulting in variability of low-temperature properties and low physical properties in larger size forgings. Creep resistance was indicated to be relatively free of such an effect. Good-quality forging can be made from the alloy in the large sizes required for land and marine turbines.

The user with the most experience, Metropolitan-Vickers, appeared well-satisfied with Mo-V steel, and they use it whenever its oxidation resistance and hardenability will permit. They are giving consideration to protection of the surface against oxidation to permit its use at higher temperatures.

Apparently 90,000-psi yield strengths cannot be obtained in aircraft disks at room temperature while retaining good creep strength. This requires higher carbon content than 0.15 percent or adjustments of heat treatment, either of which is detrimental to creep strength. While the ratio of vanadium to carbon is apparently quite critical for creep resistance, no information was obtained which suggested that increasing both carbon and vanadium would improve the alloy.

<u>1Cr-0.75Mo steel</u>.- De Havilland originally adopted this steel for early model Goblin disks because it has outstanding forgeability and hardenability for disks. Low creep strength, combined with a rapid loss in physical properties with increasing temperature of use, has limited its applicability.

<u>3Cr-Mo-W-V steel</u>.- The 3Cr-Mo-W-V steel for disks is satisfactorily used by De Havilland and they see no need of going to higher-alloyed steels at present operating temperatures. Some concern over its corrosion resistance and experimentation with corrosion protection was evident; nickel plating has been found effective for protection on present disks.

Creep requirements for disks are that the total deformation in 300 hours shall not exceed 0.5 percent at 1110° F (600° C) under 24,640 psi. Occasional spot checks show no difficulty with this requirement.

Initial difficulties with quality and variable creep resistance have been cleared up. Good creep resistance is dependent on a strict control of vanadium-to-carbon ratio to at least a minimum of 4. It was indicated that further improvement in creep resistance might be obtained by reducing the carbon from the present 0.2 percent to 0.15 percent while still retaining satisfactory yield strengths. The presence of chromium sets up a partition of carbon between chromium, molybdenum, and vanadium. To obtain best creep resistance, it is necessary to use high hardening temperatures in the presence of chromium, one effect being to control the type of carbides formed. At optimum ratios of vanadium to carbon, high-temperature properties were stated to be relatively independent of cooling rate. Secondary hardening during tempering contributes substantially to yield strength.

There appear to be two controversial points about the 3Cr-Mo-W-V steel. The necessity for tungsten is not clear and is doubtful. The high-temperature normalize before hardening is difficult to justify. Apparently, however, a beneficial effect can be shown in improved properties from the normalize when properties of a number of production disks are checked.

12Cr steels.- The amount of service experience with this type of steel is as yet limited. Disks of R.ex 448 alloy have, however, been shown to have properties equal to or better than those of the lower-alloyed steels while retaining high strength at temperatures above 1100° F. An outstanding characteristic is high ductility, particularly reduction of area in rupture tests. The better corrosion resistance is also a good feature. Apparently, the producers had pretty well solved production problems, although it was indicated that both quality and uniformity of properties had initially been a severe problem.

G18B alloy.- The interest in this alloy was not very great. Hints of trouble with disks of this alloy in the older engines due to poor quality were occasionally encountered. In general, its record seemed good. Two failures associated with very low center ductility were reported. The cause of failure apparently was high thermal stresses developed during rapid acceleration, causing fracture of center material which had poor ductility.

Facilities for warm-working G18B are of limited capacity. Large disks are warm-worked only in the center. Bristol indicated that this provided a proof strength of only 62,000 psi, and it was necessary to cold-work by overspeeding to obtain the 72,000-psi strength required in one case.

DISKS FOR LAND AND MARINE TUBINES

The practice for large turbines is not standardized. Very little experience is yet available, because the first largely experimental turbines are only in the process of being built. The usual practice has been to use the most creep-resistant alloy available in the form of the large forgings necessary for disks. This has limited the available alloys largely to ferritic and 18-12+Cb steels, although parts of large rotors have been made from R.ex 326 alloy. Mo-V steel has been utilized by the aid of cooling.

The general feeling is that the operating conditions, particularly temperatures in large turbines, are as yet unknown. Actual experience from turbine operation will have to be obtained before a better basis for selection of disk materials and disk design will be available. This is particularly true for the evaluation of welds and rotors made by a welding assembly of disks.

The general practice is to use single forging rotors insofar as possible. C. A. Parsons could not obtain a large enough forging of a sufficiently heat-resistant alloy for their largest turbine. They compromised by making two forgings and welding them together. Welds were radiographed using radiation from radioactive iridium made available by atomic energy work in Britain. They were enthusiastic about this method. Relief between rotors permitted introduction of the radio-active source. The high-temperature end was made from R.ex 326 alloy and the low-temperature end from 18-12+Cb. One company rejected a large rotor made by welding a number of Mo-V disks. Smaller turbines generally use single forgings usually of 18-12+Cb steel. No large forgings of G18B alloy or the 3Cr-Mo-W-V steel were encountered.

Inspection of the large forged rotors is difficult. Very little in the way of specification of properties can be done. Major emphasis is placed on the problem of obtaining sound forgings and whatever properties result are accepted. General experience seemed to indicate that test-sample properties, particularly creep resistance, were equal to or better than those anticipated from tests on bar stock. The usual practice is to bore out the centers of the disks and to inspect visually for flaws. Metal is removed until all poor-quality center material disappears. If flaws do not disappear before the hole becomes too large for safety, the rotors are rejected. At one company overspeed testing similar to that used in their steam-turbine practice is being contemplated.

Welding is a major problem in disk construction. Details of methods of obtaining satisfactory welds were not obtained. The subject had received considerable attention and several views were expressed on the subject.

COMBUSTION CHAMBERS AND TAIL PIPES

The materials used for flame tubes in combustion chambers are about the same as were in use 4 years ago. Designs have been changed somewhat, but Nimonic 75 is still the alloy in greatest use. When properly designed, flame tubes of Nimonic 75 run 500 to 1000 hours without replacements.

Most engines use multiple flame tubes of the "can" type. One engine uses an annular type of combustion chamber of a "vaporizer" design made from Nimonic 75, and another engine produced by this same manufacturer uses can-type combustors, but also of the vaporizer design. In this latter case, an alloy containing 18 percent chromium, 37 percent nickel, 2 percent silicon, and the balance iron is used since, in flame tubes of this particular design in which localized areas of reducing conditions are obtained, sulfur-containing fuels have corroded the combustor. Surface treatments of mild steel for flame tubes have not been successful. One producer is seriously considering the use of a ceramic coating on Nimonic 75 flame tubes similar to the coating used by Pratt & Whitney in their Nene production.

Some engine producers used outer casings for combustors made from aluminum-coated mild steel, the coating usually being produced by aluminum spraying. One producer uses nickel-plate mild steel. When a better material is required for outer casings, titanium-stabilized 18Cr-8Ni stainless steel is used.

For intermediate-temperature applications, such as for tail pipes, titanium-bearing 23Cr-18Ni steel sheet is used. Lower temperature operating components are made from 18Cr-8Ni-Ti steel.

COMPRESSOR ROTATING BLADES

The aluminum alloys RR-57 and RR-58 are reported to give satisfactory service in the low-pressure stages of axial-flow compressors.

There is no difficulty now with corrosion of the blades. Previously, some intergranular corrosion was obtained in compressor blades of RR-56 alloy, and its use has now been discontinued. In some applications, anodized coatings are used on the aluminum-alloy blades, and, in one case, a silicone coating is used in addition. Some RR-57 blades are precision stamped almost to size, and finishing requires removal of only about 0.003 inch of material.

For high-pressure stages in axial-flow compressors, one engine producer uses aluminum-bronze blades and another S-80 steel blades (20Cr, 2Ni). Still another engine producer uses forged H.R. Crown Max for all rotating compressor blades. Some cast H.R. Crown Max blades are also being considered. This strong alloy is being used by this one producer because of the necessity of having blading materials sufficiently strong to chew up ice under icing conditions. The other turbine producers have not reported any lack of ability in this regard for their aluminum-alloy and aluminum-bronze blading.

Rolls-Royce are convinced that titanium alloys will make a good compressor blade material, and a supply of the alloy has been ordered to run the necessary tests.

The centrifugal compressor in the De Havilland engine satisfactorily uses an RR-58 aluminum-alloy forging.

In the power gas turbines, rotating compressor blading is mostly either of aluminum bronze or the S-80 straight chromium (16-20Cr, 2Ni) alloy.

A railroad turbine uses 12-14Cr stainless steel. One power turbine producer prefers compressor blading for marine applications made from alloy S-80, principally because of its better corrosion resistance and better fatigue resistance, particularly under corrosive conditions, as compared with 12-14Cr stainless steel.

None of the temperatures in these compressors with less than 6 to 1 compression ratio are too high for aluminum bronze. Cost will probably be the deciding factor determining the choice between aluminum bronze or stainless iron for compressor blading. A method of precision forging of aluminum-bronze compressor blades has been worked out by High Duty Alloys Company. There is no objectionable oxidation occurring even at the forging temperatures. Some aluminum-bronze blades, 8 inches long by 3 inches wide, were in process in one plant.

These stronger and more corrosion-resistant materials, as compared with aluminum alloys, are an obvious choice for applications involving long service times where stability of dimensions and corrosion resistance are required. However, the use of RR-57 aluminum alloy for compressor blading in other than marine applications has been reported by Metropolitan-Vickers.

COMPRESSOR STATOR BLADES

The information obtained on only two aircraft engines indicated no aluminum alloys were being used as stator blades. (Note, the De Havilland centrifugal compressor does not use stator blades.) At Bristol cast H.R. Crown Max is used, which gives both good erosion and corrosion resistance. Other materials being used were aluminum bronze, cast Inconel, and cast S-80 alloy. In some instances, these latter materials were used more from the standpoint of trying out different alloy materials rather than of having been determined as optimum for the locations at which they were used. Pametrada stator blading was of an aluminum bronze containing 10 percent aluminum, 4 percent nickel, and 4 percent iron.

CERAMICS AND CERAMIC COATINGS

The development of ceramics for blading in turbines had apparently been abandoned. The probability of discovering a material of this type with adequate resistance to stress concentration and mechanical shock is considered very remote.

One group hopes that ceramics can be used for turbine stator blades and has been experimenting with ceramic blades. Compositional changes are being attempted to improve the thermal-shock resistance which apparently is not adequate.

Considerable development work was in progress at several places on ceramic coatings. Except in one instance, this was directed toward heat insulation rather than corrosion protection. The ability of a coating to reflect heat and reduce rate of heat transfer was being considered for use in reducing thermal shock, particularly in combustion chambers and stator blading. In addition, any cooled part would be aided by a coating which reduced heat-transfer rates. One manufacturer was working on a ceramic coating for blades which would protect against lead corrosion in the event that it became necessary to use leaded gasoline for fuel.

CERAMALS

The most important ceramal development appeared to be a ceramal of 40 to 50 percent titanium carbide, 5 to 10 percent chromium carbide, and the balance nickel (or cobalt) by Hard Metal Tools, Ltd., Coventry, in conjunction with Armstrong Siddeley. This material had been used with encouraging results for blades in an engine.

The operating temperature in this engine was only 1380° F (750° C). On the basis of this development, there was considerable interest in the possibility of substituting ceramal for metal blades for operation in the temperature range 1380° to 1650° F (750° to 900° C). Advantages of the ceramal blades would be better thermal-shock resistance than that of currently used metals, lower stresses due to lower density, higher modulus, better creep resistance and equal fatigue resistance to metals, and a relatively inexpensive, good, high-production process economical of strategic materials.

A cobalt-base ceramal of the type indicated above gives better strength at high temperatures and possibly better quality bodies, but work was recessed because cobalt is considered to be too scarce. The nickel-base ceramals are easier to fabricate, particularly in grinding, than the cobalt-base material.

One interesting and puzzling development was that chromium completely penetrated Hard Metal Tools ceramals during chromizing. The penetration was much faster than is normal for metals.

The disadvantages of the ceramal blades are mainly susceptibility to cracking as a result of stress concentrations produced in fastening them to wheels and low mechanical shock resistance. Most engine company representatives were rather pessimistic about the possibility of successful operation unless the low mechanical shock resistance could be overcome in the material. One report of oxidation resistance for this ceramal lower than for Nimonic 80 was encountered, although Armstrong Siddeley reported it to be superior in this respect to Nimonic 80 in the engine tests.

Work on Cr-Al_2O_3 ceramals is being sponsored by the National Gas Turbine Establishment with, as yet, inconclusive results. There is a possibility of considerable industrial development in this field beyond that observed.

In spite of the difficulties to be overcome in the use of ceramals, some metallurgists in Britain feel that the next new material for gas-turbine service will be the titanium carbide-metal ceramals. Differing from current United States thinking on the use of these materials, many feel that these ceramals should be used first in the present operating temperature range of 1375° to 1475° F. Their lower density would permit higher rotational speeds and greater engine output with no greater blade stresses than are now being used.

SURFACE PROTECTION WITH DEPOSITED METALS

Aluminized low-carbon steel was widely used for the outer casings of combustion chambers. Nickel plating was being used in one case for the same purpose. Nickel and chromium plating of miscellaneous parts was observed to some extent. Most aircraft companies were experimenting with, but not yet using chromized parts. One turbine laboratory was aluminizing austenitic 18Cr-8Ni steel bolts by spray coating, followed by diffusion at 1470° to 1560° F (800° to 850° C) for 15 minutes to prevent seizing after high-temperature exposure. Chromizing of the nuts is almost as beneficial.

Chromizing is, as yet, on a small scale and largely in a developmental state. Some use is being made of chromizing for miscellaneous parts in the chemical and corrosion fields. Two companies, The Metal-Gas Company and the B.S.A. Group, are vigorously promoting its use. Chromium-alloy layers about 0.020 inch thick are produced by deposition and simultaneous diffusion of chromium from atmospheres containing gaseous chromium chloride. The chromium content at the surface is 40 to 50 percent. When properly produced, these chromized coatings can be rolled or processed hot or cold without destroying the coating.

The chromizing process is being developed chiefly for the protection of ferritic steels with good strength, but inadequate corrosion resistance at high temperatures, particularly in tube form. One severe handicap of chromizing is the tendency for carbon to build up in the coating by diffusion from the steel during deposition and service. Carbon build-up results in a hard and brittle surface. Only special low-carbon steels such as Armco iron have ductile surfaces. The method being used to avoid this difficulty was to develop special steels free of carbon embrittlement. Mo-V steel free of this effect and having superior creep resistance up to 1200° F with otherwise suitable properties for gas-turbine disks was indicated to be a distinct possibility.

One laboratory showed results in which the break in the stress-rupture-time curve for Mo-V steel, usually associated with intergranular oxidation, was eliminated by protection by chromizing, thereby considerably improving the long-time rupture strength.

TITANIUM-, CHROMIUM-, VANADIUM-, AND MOLYBDENUM-BASE MATERIALS

Titanium

The work on titanium in Britain has been very limited in scope, principally because of lack of titanium metal with which to work. The Royal Aircraft Establishment made a small amount of metal using the Kroll process, but terminated their production because it seemed more properly an industry function to supply metal for research at various laboratories.

Kroll-process sponge has been supplied by Brown-Firth under a Ministry of Supply contract, but production has been only about 300 pounds in the past year. A typical analysis of their sponge is as follows: 0.2Fe, 0.004Mn, $0.17O_2$, 0.35Mg, 0.05C, $0.05N_2$. These results indicate a composition somewhat less pure than American produced sponge.

The Imperial Chemical Industries, Ltd., are going to make titanium by the Kroll process on a prototype scale. This large chemical company will probably be the producer of titanium sponge in Britain.

Dr. Gross of Fulmer Research Institute is working on a monohalide distillation process for titanium production and also is appraising the potentiality of an electrolytic process.

In work at the British Non-Ferrous Metals Research Association, metal melted from Brown-Firth sponge showed tensile strengths ranging from 112,000 to 123,000 psi and with elongations of about 22 percent and reductions of area of 30 to 35 percent. In induction melts made in carbon crucibles, carbon contents of 0.5 to 0.6 percent were obtained and are not considered objectionable.

In their work on alloys of titanium, the effects of a wide range of additions of many different elements were studied. Additions of 1, 5, 10, 25, and 50 percent were tried, using powder-metallurgy techniques. Later studies were restricted to alloys which could be cold-rolled, and these included binary alloys with 10-percent additions of molybdenum, columbium, and zirconium; with 5-percent additions of chromium, aluminum, tungsten, iron, and nickel; and with 1 percent of bismuth. Those additions which contributed to improved oxidation resistance were aluminum, silicon, zirconium, and beryllium.

There was some trouble in rolling titanium alloys, but by preheating the rolls to 300° F (150° C), rolling was conducted at a metal temperature of 1380° F (750° C) without trouble. Rupture tests at 750° and 840° F (400° and 450° C) are just beginning on carbon-melted sponge titanium and on several binary alloys using aluminum, tungsten, zirconium, silicon, and chromium additions. The program of the British Non-Ferrous Metals Research Association is not extensive and, with research in the United States so far ahead, they may cut back or entirely eliminate their work on titanium.

The National Physical Laboratory initiated a program recently to determine the titanium-carbon-oxygen system. Van Arkel process titanium will be used. The work is just starting. They are working out methods using less-pure titanium.

As a result of their thermodynamic studies of titanium reduction processes, they belive that titanium tetrachloride can be reduced at very high temperatures by hydrogen, but have not been able to do so. Their studies of the possibilities of electrolytic reduction processes for the production of titanium metal have not developed any compounds suitable for electrolysis.

Chromium

Fulmer Research Institute showed little optimism about the probability of producing chromium-base alloys of adequate ductility and consistency of properties. Their work has indicated that 10 percent of iron in chromium is the optimum matrix for alloying additions. Using tantalum, columbium, molybdenum, tungsten, and vanadium, 10-percent additions were indicated optimum, but the high-temperature properties of the alloys at 1600° F were not so high as for the alloy containing 60 percent chromium, 15 percent iron, and 25 percent molybdenum reported by the Battelle Memorial Institute.

Chromium metal made by powder-metallurgy techniques using powder from chromium hydride decomposition has been processed by Fulmer. Chromium metal so produced shows a sharp transition from brittle to ductile behavior in bending at 465° to 480° F (240° to 250° C). Such metal can be successfully hot-worked at 930° to 1650° F (500° to 900° C).

At the National Physical Laboratory it has not been possible to produce pure chromium metal which is ductile. They know of no fundamental reason why chromium should not be ductile - although there may be an as yet unidentified crystal structure present in pure chromium.

Vanadium

The properties of a large number of pure metals have been studied by the National Physical Laboratory and compression creep strength at 1830° F (1000° C) has been determined for the chromium, molybdenum, and tungsten group and the vanadium, columbium, and tantalum group. As a result of Dr. N. P. Allen's last visit to the United States, vanadium is considered to be of most likely interest in the vanadium, columbium, and tantalum group and also available in usable quantities at the least prohibitive cost. They, therefore, undertook the study of vanadium in greater detail. Columbium had very interesting properties, but its scarcity has caused NPL to drop its development.

Vanadium is not oxidation resistant but has reasonable ductility with relatively high impurity content. Material with 0.4 percent oxygen is quite ductile. Properties at high temperatures were interesting, but not so much so as to displace molybdenum and chromium in their work. Their work has been done on vanadium powder prepared in Britain, with purity of the order of 99.8 percent. The NPL showed interest in the Union Carbide and Carbon Corporation development of vanadium metal and their offer of metal for sale, and may do additional laboratory work on vanadium.

Molybdenum

Conversations in Britain at the laboratories visited did not disclose much information on work on molybdenum there. Metallurgists were very interested in American work on molybdenum, and much more information was given on this subject than was reveived.

Wm. Jessop and Sons have obtained a bar of arc-cast molybdenum from the Climax Molybdenum Company and are beginning to study its creep properties, welding characteristics, and possibilities of protection by chromizing. It is suspected that considerable work had been done on chromizing of molybdenum, but no details as to the process or its effectiveness were offered. Wm. Jessop and Sons have also done some powder-technique work on molybdenum, so surely some molybdenum-alloy work has been done also. They are interested in this work because of the need for harder molybdenum-rich gun-tube-liner alloys. It was indicated that they could weld molybdenum, but no details could be obtained.

Armament research is the principal concern of the Defense Weapons Research Department of Woolwich Arsenal and, in this connection, they are working on molybdenum and its alloys for gun liners.

In an erosion-testing gun, a molybdenum liner cracked after firing two rounds, but no weight loss was shown.

This work on molybdenum is just beginning at Woolwich Arsenal. A new vacuum-arc-melting furnace has been put into use. It can melt up to 50 pounds in the form of a 3-inch-diameter ingot. Melting is by means of a consumable electrode of sintered molybdenum purchased from Murex Welding Processes, Ltd. Carbon deoxidation is used.

In their alloy work which is just beginning, it is planned to investigate solid-solution alloys first. Evaluation of forged material will be by hardness, and later by hot tensile and creep tests. Forgeability will be evaluated by upsetting tests, but they think extrusion may be a better method. The optimum temperature for upsetting molybdenum was indicated to be 2125° to 2200° F (1170° to 1200° C).

Since guns under rapid-fire conditions sometimes have their bores brought up to a red heat, experiment with protective coatings for molybdenum are being made at the Arsenal. Molybdenum wires are packed in metal powders, carbides, silicides, or borides, and a coating is diffused on by heating in an atmosphere of hydrogen and hydrochloric acid gas. In their tests to data, zircomium compounds have given the best protection, better than silicon-containing coatings. Mixed silicon and boron coatings are better than silicon alone at 1830° F (1000° C), but not at higher temperatures. Some nickel-chromium alloy coatings are being tried with the hope that such a coating may be more perfect and also more ductile than a silicon or silicide coating.

STEAM PIPING

Six different grades of steel are being used for steam piping in Britain. These are (1) low-carbon steel, (2) 0.5Mo steel, (3) 1Cr-0.5Mo steel, (4) 2.25Cr-1Mo steel, (5) 0.50Mo-0.25V steel, and (6) 18Cr-8Ni-Cb stainless steel.

Present British practice is to make low-carbon-steel main-steam-line piping with an aluminum addition of 2 pounds per ton and limit its use to 750° F maximum. In the current British specification for heavy superheater tubes intended for about 3 years' service, a steel is required to be tested at 842° F (450° C) at 17,920 psi for 2 days, and the minimum rate should not be in excess of 50×10^{-6} inch per inch per hour. Such a creep rate would be too high for a steam-piping material for long service life, and, in this case, a creep rate of about 5×10^{-6} inch per inch per hour would be required.

Both The United Steel Companies and the National Physical Laboratory are studying the effect of deoxidation practice on the creep properties of plain-carbon steels. The research of the former company has indicated that, if the soluble aluminum exceeds about 0.015 percent, the creep rate in the 24- to the 48-hour period is almost always greater than 50×10^{-6} inch per inch per hour, which is indicative of poor creep resistance for the steel. At the NPL, on the other hand, evidence has shown that the amount of soluble aluminum is no certain indication whether a steel will be adequately creep resistant.

To meet best the above-noted specification, silicon-killed steels with no aluminum or up to 1 pound of aluminum per ton are used.

For the temperature range of 750° to 950° F, C-Mo and the 1Cr-0.5Mo and 2.25Cr-1Mo steels are used.

An extensive program was in progress at The United Steel Companies on the rupture properties of bars with three Izod notches in the gage section. Preliminary results showed only one-fifth to one-eighth the life of unnotched specimens for 0.5Mo steel tested at 1050° F (stress based on area at base of notch). The notches greatly increased the scatter in data, ranges of 100 percent being observed. They indicated that the loss in rupture life as a result of the notches was markedly affected by steel-making practice, low aluminum causing sensitivity to notches. The observation was made that the lower-carbon steels generally made in the United States tend to show a much greater scatter than British higher-carbon steels.

Four power stations are being constructed in Britain for a steam temperature of 1050° F. Three are using 18Cr-8Ni-Cb piping, and the fourth is using 2.25Cr-1Mo steel. Some feel that the Mo-V steel is superior to 2.25Cr-1Mo in creep strength, fabrication characteristics, and weldability, when suitably heat-treated by normalizing from 1750° to 1800° F, followed by tempering at 1275° F. Both the high-temperature solution treatment and the tempering treatment must be used to produce the desired precipitation and strength effects. The full beneficial effect of heat treatment and tempering is not realized nor indicated in short creep tests of 1000 hours' duration. Metropolitan-Vickers plan to use Mo-V steel for steam piping at temperatures up to 1050° F, since they feel that the 2.25Cr-1Mo steel shows lower creep resistance at 1100° F than the Mo-V steel.

A temperature of 1100° F is regarded as borderline for the Mo-V steel with regard to oxidation resistance. For service at temperatures up to 1140° F using Mo-V steel (now planned by some engineers), protection against oxidation by chromizing will be used.

A ferritic steel still better than this Mo-V steel is needed. Such a steel would have equal creep strength at somewhat higher temperatures and also possess better oxidation resistance.

BRAZING AND WELDING

Considerable work has been done on the development of high-temperature brazing materials. The silver-palladium-manganese alloys developed by The Mond Nickel Company are the best found, so far, showing quite useful strengths as high as 1200° F (650° C). These range from alloys containing about 75 percent silver, 20 percent palladium, and 5 percent manganese, up to alloys containing about 80 percent palladium, 5 percent aluminum, 5 percent manganese, and 10 percent silver. The better alloys, available in powder form only, braze at temperatures as high as 1975° F (1080° C). One alloy shows a shear strength at 1470° F of about 13,000 psi. It will support a useful load of 2240 psi at 1110° F (600° C).

Titanium-containing alloys can be brazed without a flux in hydrogen, but the Nimonic alloys containing both aluminum and titanium tarnish and cannot be braxed without a flux.

Colmonoy No. 6 also brazes quite well, but it does show some corrosive attack by intergranular penetration of Nimonic alloys. The silver-palladium-manganese alloys do not show this intergranular attack.

There was considerable discussion on welding of high-strength alloys. Very little information was available on the strength and structural stability of such welds at high temperatures.

Metropolitan-Vickers fabricated some large rotors by welding together a number of small disks of Mo-V steel. The welds were made with austenitic weld rod, but initial experiments with an 18Cr-8Ni rod produced sigma phase in the welds on stress relieving. A more highly alloyed austenitic rod, R.ex 326 (0.25C, 19Cr, 15Ni, 3Mo, 7Co, 2Cb), gave satisfactory welds. Metropolitan-Vickers would like a ferritic weld rod, but no satisfactory one is as yet available.

Austenitic welds with not over 5 percent of ferrite are advocated by Brown-Firth. Rods are used by C. A. Parsons which, under the welding conditions used, give just under 5 percent ferrite in the deposits. The problem is to have enough ferrite to prevent cracking during welding but not enough to cause trouble from sigma phase during service.

In discussions with Wm. Jessop and Sons on cracking in austenitic weld metals, it was stated that any columbium in excess of the quantity required to stabilize the carbon can react with silicon and iron to form a ternary intermetallic compound, an iron-columbium silicide. The compound $Fe_4Cb_5Si_3$ is one of three of this type, and when it occurs as a massive constituent in the grain boundaries, it has been associated with weld cracks. Their work suggests reducing the columbium and silicon contents. Top ratio of columbium to carbon should be 13:1, since the theoretical ratio is about 8:1. It is not easy to regulate the columbium-carbon ratio in commercial practice; it seems more easy to reduce the silicon content.

RESIDUAL FUEL-OIL ASH CORROSION

The corrosion of gas-turbine blade alloys by ash from combustion of residual fuel oils is being studied by many laboratories in Britain. The future of the power gas turbine appears to be dependent on a successful solution of the corrosion problems resulting from the vanadium pentoxide (V_2O_5) in the ash. One disturbing influence on the future of gas turbines for marine power is the successful use recently of residual fuel in marine Diesel engines. The entire Anglo-Persian tanker fleet is being converted to burn residual oil in its Diesel engines after a several-year trial in a few ships of the fleet.

Because it is molten at desired gas-turbine operating temperatures, the V_2O_5 adheres to nozzle vanes and blades, fluxes and attacks the protective oxide film on the alloys, and then chemically attacks the base metal.

In early gas-turbine experiments, blocking of nozzle blade channels with V_2O_5-containing ash occurred. This difficulty was eliminated by debasing the combustion and forming carbon; the carbon particles absorbed and covered the V_2O_5, thus reducing both adherence and corrosion attack. Corrosion seems to be proportional to the amount of deposit that sticks to the blades.

Although considerable effort has been devoted to the problem, no practical solution to the V_2O_5 corrosion problem was offered by any of the laboratories. Nickel and cobalt appear to be the metals most resistant to attack; therefore, nickel- and cobalt-base alloys are more resistant than the chromium-nickel-iron heat-resistant alloys. Nickel oxide does not form so low a melting-point slag with V_2O_5 as does iron oxide. This is important since a primary requisite for resistance to V_2O_5 corrosion is the avoidance of a liquid slag. Nickel plate appears to help in protecting some materials.

One laboratory suggested that V_2O_5 forms vanadates when it attacks protective metal oxide films. Experiments were conducted in which powdered vanadates were added to a 1 percent solution of Canada balsam in xylene, and then painted on metal surfaces which were then heated at gas-turbine operating temperature. Nickel and calcium vanadates do not promote attack. Magnesium and zinc appear promising in preliminary experiments. Chromium vanadate does attack metals, and the attack is proportional to the V_2O_5 in the chromium vanadate. The attack of V_2O_5 on both cobalt-base and nickel-base alloys seems to be proportional to their chromium contents.

Nickel in contact with V_2O_5 at 1470° F (800° C) starts to corrode, but then the corrosion rate falls off to about the oxidation rate in air at that temperature. Aluminum is about as bad as chromium in promoting attack by V_2O_5. Molybdenum was generally reported to be very detrimental, and iron was not much better.

No alloys presently available are considered to have satisfactory resistance to V_2O_5 . The nickel-base alloys are best but still are not good enough for long-time service at 1290° F or above. They may be adequate for 1000-hour service if the temperature is not too high. A cobalt-base alloy of the H.S. 21 type, containing 6 percent molybdenum, showed about the same amount of attack at 1470° F (800° C) as the Nimonic alloys.

All the common ceramics are no better than (if as good as) the metals. The titanium-carbide-nickel-chromium ceramal made by Hard Metal Tools, Ltd, is rapidly attacked at 1380° F (750° C) and 1470° F (800° C).

The limited experience to date with power-turbine operation burning residual fuels suggests that the corrosion in the turbine may not be nearly so severe as would be indicated by many of the laboratory tests. Many experimenters do not believe that estimates of resistance to V_2O_5 corrosion obtained by exposing metal surfaces to large quantities of V_2O_5 are any indication of service behavior in a turbine.

Considerable work has been done on addition agents to the oil, but nothing of promise was reported. Greater corrosion was reported when vanadium was artificially introduced into vanadium-free oil as compared with oil with naturally occurring vanadium content.

Other than the possibility that corrosion in the turbine may not be so severe as indicated by some laboratory tests, the only item of promise appears to be control of the combustion so as to produce carbon to absorb or coat the V_2O_5 and prevent its adherence to the blades. It was even suggested that additions be made to the oil to promote the desired type of carbon formation during combustion.

FATIGUE AT HIGH TEMPERATURES

The usual attitude of engine producers was that fatigue problems are solved by design changes. In general, alloys were considered to be limited in use by creep rather than by fatigue resistance in any application where creep occurs to a significant extent. Many engineers feel that fatigue properties are more important when compared on a time basis. In combined stress tests, the mean stress is important, because it has an effect on a time basis. Only one group of design engineers was found to be taking superimposed dynamic loads into consideration.

Wrought materials were generally considered to have sufficient fatigue resistance so that probable performance of such materials is judged mainly on creep resistance. Cast materials were condemned for moving blades on the basis of unpredictable fatigue resistance. Castings were not considered inherently too low in fatigue resistance, but it was considered impossible to eliminate the flaws in all castings which lead to low fatigue strength.

All turbine engineers felt that fatigue strength could be expected to increase with cyclic speed, yet there was a universal desire to have the ultra-high-frequency ranges encountered in turbines investigated. Those building long-life turbines were considerably more concerned about fatigue effects than the jet-engine engineers.

One view was expressed, with some supporting evidence, that fatigue failures will only be encountered, in turbines free from resonant vibrations, after third-stage creep had started. Occasional fatigue failures in blades were claimed to be associated with those blades which had low creep strength and entered third-stage creep prematurely.

The available information indicated that bending tests should give about the same results as axial tests in fatigue, provided the stresses are accurately known and the temperatures properly measured. Any difference observed would be small and due to the greater probability of encountering flaws in the large stressed sections of axial-test specimens. The axial tests should give slightly lower results, although the effect should decrease with increasing temperature because of relief of the stress concentrations at flaws. In some tests at Rolls-Royce, up to 1650° F, fatigue strengths from tests in the Haigh Machine were about 30 percent lower than those in rotating bending.

Conflicting claims were made regarding the reliability of various testing machines. For bending tests, there seem to be some questions regarding the stress calculations. Reported values tend to be high, because stress-redistribution effects are not taken into consideration. The axial tests were criticized for overheating during application of the test loads and the difficulty of obtaining pure axial load without bending. Temperature measurements in rotating-beam machines were frequently doubted. In all cases, there was likewise staunch support for the reliability of each type machine. As eminent an authority as Dr. H. J. Tapsell of the National Physical Laboratory, however, preferred axial-type machines because of more certainty of stress. One engine producer was testing blade shapes in a combined steady tension and bending machine operating at about 700 cycles per minute.

Notched fatigue specimens (V-notch) show big reductions in fatigue strength at room temperature. At 1500° F, on Nimonic alloys, notches show very little effect on fatigue strength and the transition temperature above which the effect of the notch decreases rapidly is about 1300° F (700° C).

No systematic information on the effect of surface finish was found. Cases where severely cold-worked surfaces on Nimonic alloy blades led to fatigue failures were reported. In this case, the real explanation appeared to be low creep-rupture strength resulting from residual cold-work and precipitation leading to intergranular cracking which, in turn, caused low fatigue strength. Otherwise, general experience indicated that the common surface-finish effects decreased with increasing temperature so that, at temperatures where creep occurs, fatigue properties are relatively insensitive to surface finish.

FUNDAMENTAL RESEARCH ON CREEP AND FATIGUE

Research on the fundamentals of creep was widespread. Industrial laboratories, as well as Government laboratories and educational institutions, were devoting large parts of their efforts to this problem. Fundamental work on fatigue was also being actively investigated, but not nearly to the extent of creep in those establishments visited.

Quite separate objectives confine the work, in most cases, to the mechanism of the creep or fatigue processes, metallurgical effects, or stress considerations. While a complete understanding of creep and fatigue at high temperatures is still far from being attained, the prospects appear to be quite promising.

Polygonization concepts were occupying the attention of a number of investigators of the creep process. Dr. W. A. Wood's original results have been generally verified, although most experimenters do not agree that creep occurs by the movement of the individual cells. Results indicate that polygonized structures can be produced by prior treatments. Those resulting from hot-working are different from those resulting from cold-working or from the creep process itself. Apparently, the creep resistance is higher the smaller the size of the cells produced by prior treatment. The polygonite boundaries tend to behave like grain boundaries at high relative temperatures. Present techniques indicate that precipitates prevent polygonization, although this may be due to the limited sensitivity of present methods. Apparently, polygonization effects have been explained in terms of dislocation theory.

Polygonites in fatigue specimens of aluminum, copper, and "pure" iron have been observed by the Royal Aircraft Establishment. Metallographic techniques indicate that they originate at the interface between deformation bands and the rest of the grains.

Studies of the relative movement of grain boundaries and grains during creep are in progress. The usual view is that grain boundaries move first under an applied load, setting up stress concentrations within grains. This causes slip within grains which, in turn, allows further grain-boundary movement. Thus, the factors controlling the creep process can be visualized in a simplified manner as grain-boundary "viscosity," the resistance to slip of the matrix, and the rate of relaxation of internal stress within the grains. The relationship between this concept and the concept involving polygonization effects is not at all clear, because the latter effect is essentially a matrix phenomenon. A simultaneous investigation of measurement of relative creep in grains and grain boundaries in polygonized structures at the National Physical Laboratory is producing very interesting results.

Most investigators feel that the grain-boundary conditions and the grain size are extremely important to the creep process. Indications are that the importance of grain boundaries increases with test temperatures and eventually a temperature is reached at which virtually all creep occurs by movement along the grain boundaries. Where grain-size variations have been studied, the effects on creep are small in comparison with those possible from other metallurgical variables and in no case are free from the possibility of other effects being the controlling factor. It appears that most fundamental investigations have concentrated on temperature ranges where most creep takes place at grain boundaries. Thus, application to the relative temperatures at which engineering materials are used, where movement within the grains appears to be a substantial part of creep, is a considerable way off. The investigations in this field at the National Physical Laboratory, the Fulmer Research Institute, and the Royal Aircraft Establishment for the Ministry of Supply and the internal work by The Mond Nickel Company are excellent.

The work on metallurgical variables affecting creep and fatigue was most interesting. While the pure scientists are just starting in this field, it appears that their results are going to be more in line with the views held by metallurgists dealing with engineering materials. The few data obtained to data regarding presence of alloying elements, mainly at the British Non-Ferrous Metals Research Association, indicate that solid-solution effects are the controlling factor and, when precipitates form, they may completely reverse grain-size trends. It even appeared that such laboratories as those of the Physics Department of Bristol University are tending to use alloys rather than "pure" metals for this reason. Metallurgists at commercial organizations, with one exception, generally reported that solid-solution or precipitate effects within grains seem to be controlling factors. Laboratories at commercial companies have, in general, concentrated on establishment of phase diagrams for the complicated systems of alloys with which they deal. This, combined with analysis of excess phases obtained by chemical separations, appears to have provided a basis for development of better alloys and for control of the properties of those already made. The tendency appears to have been to approach the problem from that angle, rather than to devise means of measuring the effects of the solid solution and excess phases on the internal structural conditions of the alloys. The work of the Wm. Jessop, Brown-Firth, and Mond Nickel Company laboratories has been excellent and successful in this field.

Both the pure scientists and the practical metallurgists are quite conscious of the tendency for hot-worked materials to have high creep resistance. The NPL metallurgical group is particularly interested in this finding for materials as diversified as pure aluminum and carbon steel. In the case of pure aluminum, this was associated with an initial structure showing more and finer polygonization than the structures developed by other treatments.

The engineering group at NPL is continuing its fine work on creep and fatigue under complex stresses. Excellently conceived and executed programs are in progress on creep under complex stresses in nonisotropic metal, application of simple stress-creep data to complex structures, relaxation phenomena, variation of temperature and stress during tensile creep tests, speed of stress reversals in fatigue testing, and combined stresses in fatigue testing.

FUNDAMENTAL INVESTIGATIONS IN PROGRESS

1. A coordinated program of fundamental research on creep and fatigue is being sponsored by the Ministry of Supply as follows:

(a) Effects of solid solutions on creep resistance at the Royal Aircraft Establishment under Mr. Daniels. Initial work is on the "perfect" solid solution of silver in aluminum. Alloys containing 0.5, 2, 4, and 8 percent silver are to be creep tested over a range of temperatures and stresses, using the 0.5 percent alloy as a base line, because superpurity aluminum has proved too variable for this purpose. Later, the work will be extended to cover additions of solute atoms to copper and silver. Preliminary results indicate very little difference between 0.5 and 2 percent silver in aluminum. The complete data will permit calculation of activation energies.

(b) Fundamentals of fatigue at the Royal Aircraft Establishment. The results discussed consisted mainly of microscopic examination of superpurity aluminum, silver alloys of aluminum, copper, and iron. Polygonization has been observed in these materials and is associated with the initiation of fatigue failures. Deformation bands form above the fatigue limit and not below. Polygonized structures appear in the deformation bands by a sort of "recrystallization" process. The boundaries of the polygonites eventually appear to behave like grain boundaries, and fatigue cracks start in such boundaries. Increasing the temperature of testing apparently increases the ease with which polygonization occurs.

(c) Study of the mechanism of creep at the National Physical Laboratory by Mr. D. McLean. This program covers the effects of initial structures on the mechanism. Work is in progress on superpurity aluminum with various prior structures in which the movements of grains and grain boundaries during creep are measured by an interferometer technique. The fundamental interpretations are based on the dislocation theory.

Hot-worked aluminum had the highest creep strength, cold-worked intermediate, and annealed the lowest creep strength. The hot-worked material was initially polygonized to a rather fine size; cold-worked material polygonized to a somewhat coarser size either during testing or by partial annealing; and annealed material polygonized during testing to a much coarser structure. The explanation was that in hot-worked material dislocations are stabilized in rows to form the polygonite boundaries. When cold-worked, the dislocations are scattered in the lattice so as to interfere with one another. The polygonites formed by hot-working are tilted from one another by several degrees; those formed by cold-working are tilted only a fraction of a degree.

They do not believe that creep occurs by movement of polygonites, but they influence dislocations which in turn influence the slip mechanism. Polygonization is revealed by the phase-contrast microscope.

They have found for quite large creep deformations that coarse-grained material creeps 2 percent in the grain boundaries and "fine"-grained (still very coarse) about 10 percent in the grain boundaries. Even though much finer grained, commercial materials will, however, show much less creep than the material being studied, so they are uncertain about the application of their findings to such materials. The curves for grain-boundary movement have the exact shape of the creep curve, merely being displaced in magnitude. The velocity values are 100 times less than those measured by Dr. T'ing Sui Ke at the University of Chicago.

Polygonized structures obtained by hot-working maintain superior creep resistance up to the temperatures where polygonite boundaries start to behave like grain boundaries; whereupon the material then behaves as if very fine grained. Cold-worked materials maintain their creep strength up to the temperature at which dislocations are annealed out.

They are now working on copper and eventually will work on iron.

(d) Effect of insoluble phases on creep resistance at the British Non-Ferrous Metals Research Association. Iron is being added to super-purity aluminum to form an insoluble phase, the limit of solubility being 0.01 percent at 68° F (20° C) and $<$0.02 percent at 572° F (300° C). Contents of 0.01, 0.02, 0.04, and 2.4 percent iron have been tested, with the major strengthening effect obtained up to the solubility limit and with only minor increases in creep resistance beyond the solubility limit. The effect on fatigue was less and did not show a sharp transition at the solubility limit.

(e) A study of the mechanism of creep at Fulmer Research Institute under Dr. Sully. This program was an outgrowth of the observations obtained in the small-scale compression-creep unit developed for the Ministry of Supply. The compression-creep curves were similar to tension curves, and third-stage creep started at about the same time. In compression, the third-stage creep occurred as a series of steps, however.

The reasoning was that movement occurs at grain boundaries, setting up stress concentrations at the intersections of boundaries. Third-stage creep results from recovery of these stress concentrations. In tension, this leads to intergranular separation. In compression, third-stage creep is caused by relief of strain hardening as a result of recovery. This is followed by strengthening, due to the material again strain hardening, causing a stepped creep curve. Grain-boundary separation does not occur in compression, permitting such a mechanism to operate for prolonged periods.

Damping measurements similar to those developed by Dr. Ke at the University of Chicago are being used to check this theory. If fissures form, damping should increase; recovery should result in unchanged damping. The work is being done on aluminum-copper-beryllium alloy with impoverished areas adjacent to the grain boundaries by precipitation, a condition which should be favorable to recovery. Previous work on superpurity aluminum was unsuccessful, because recrystallization and necking down occurred without fissuring.

2. Grain-boundary precipitation is being studied by electron-microscopic techniques at the British Non-Ferrous Metals Research Association.

3. Electron-diffraction techniques are being developed by the British Non-Ferrous Metals Research Association.

4. A correlated program is in progress at the British Non-Ferrous Metals Research Association on the creep and fatigue, and at King's College, University of Durham, on oxidation characteristics in combustion atmospheres of hot-stamping bronzes (5-12Al, 5Ni, 5Fe). Remarkable creep strengths and amazing oxidation resistance have been obtained. Best creep strength (0.075 percent creep in 500 hr at 750° F (400° C)) under 8950 psi results from 5 to 8 percent aluminum. A minimum of 5 percent aluminum is needed for oxidation resistance. Hot-worked material is much stronger than material given any heat treatment. Additions of 6 percent aluminum produce very oxidation-resistant materials up to 1650° F. Specimens remained bright after 48 hours at 1650° F when 0.5 percent chromium and $2\frac{1}{2}$ percent aluminum were added together.

Other work is in progress, with promising results, to extend the development of aluminum bronzes for very high temperatures and prolonged service as heat-exchanger tubing.

5. A fundamental investigation of the oxidation of cobalt- and nickel-base alloys with additions of chromium, tungsten, molybdenum, and aluminum in kerosene combustion atmospheres (with excess air, as used in jet engines, air-fuel ratio 60:1) is in progress at King's College for the Ministry of Supply. Binary alloys of pure metals have been completed. The results indicated cobalt-base alloys were inferior to nickel-base alloys. A 36 percent chromium binary with cobalt had been selected as a base for further study of more complex alloys.

The work was being extended by students to cover catastrophic oxidation of alloys containing molybdenum and tungsten on the basis that Dr. M. G. Fontana's explanation in the United States was incorrect.

6. Temper-brittleness investigations were in progress at King's College, Cambridge University, and the National Physical Laboratory. Comments indicated that others were in progress in the country. The work at King's College suggested that chromium and manganese phosphides precipitated intergranularly were the cause of temper brittleness.

7. The studies of creep under combined stresses at the NPL are being continued by Mr. A. E. Johnson. The previous work on isotropic aluminum, copper, and steel, mostly published, had shown that creep curves could be made to coincide by maximum-shear-energy theory. This is being extended to isotropic Nimonic 75. The first approach to non-isotropic materials had been started by using induced anisotrophy from prestraining with preliminary results indicating only displaced creep curves of the same form.

Various types of light-alloy beams with eccentric loading are being studied. Results indicate that the creep behavior of the structure can be predicted from tensile-creep data for 0.1 percent creep, but not for 2 percent (i.e., when creep is in first stage, and possibly second stage, where their formulas apply, but not for third stage). The correlations indicate that the materials behave as if a change in modulus occurs as creep occurs, rather than as would be assumed from a stress distribution in layers inversely proportional to the distance from the neutral axis.

8. Cyclic-load and temperature effects on creep are being studied by the NPL. The schedules used simulate steam power overstressing and over-temperature operation. The results agree with the predictions of Dr. R. W. Bailey of Metropolitan-Vickers. Creep is cumulative and proportional to the time at stress and temperature. There is no recovery effect, any recovery being lost in a very short time after reapplication of load and temperature.

9. The fundamentals of recovery from creep are being studied by the NPL. Very small creep strains were being studied with the expectation that the work would be extended to larger strains. A torsion machine is being used to obtain high sensitivity.

Small strains are 100 percent recoverable. The indications are that the amount of recovery is limited by the amount of internal elastic strain induced by creep, recovery being the relief of this strain.

10. The effect of cyclic speed on fatigue is being studied by the NPL. Initial work on RR-58 aluminum alloy has been completed

at 392° F; work is in progress on low-carbon steel at 752° F, and a heat-resistant alloy is to be included.

For aluminum and steel, the fatigue strength rises to a constant value at about 200 cycles per minute. Increasing the speed reduces deviation from elastic behavior. Increasing temperature tends to increase the speed at which elastic behavior occurs. Fracture time lies somewhere between direct proportionality to speed and independence of speed - tending to be less dependent on number of cycles as speed increases and temperature decreases. The cycles for a fatigue limit increase with temperature.

Curves are being developed for 0.1 percent creep strain in combined static-dynamic tests. Location of the 0.1-percent curve is dependent on the coefficient of the equation describing the creep curve for maximum stress. Increasing the speed of reversals or the fatigue strength adds only a rather small amount to the permissible load for limited deformations under combined stress.

11. Equilibrium diagram of iron-chromium-nickel system by Dr. Reese at NPL. The definition of the sigma phase field at 1022° F is being completed. The effects of carbon, nitrogen, molybdenum, and columbium on the phase boundaries are being determined.

12. The fundamentals of the creep and fatigue of lead are being extended by the British Non-Ferrous Metals Research Association to include: (a) Complete solid solutions of tin, (b) completely insoluble phases formed by adding iron, (c) age-hardening alloys formed by antimony additions.

Creep resistance was increased by solid solution at constant grain size and decreased with decreasing grain size at constant solute content. Fatigue strength increased with the amount of solute added regardless of grain size. An insoluble phase increased creep and fatigue resistance. Both increased with a decrease in grain size due to retained cold-work. Precipitation effects have not yet been worked out.

13. At Colvilles, Ltd, an investigation is in progress which suggests that peaks in Jominy hardenability curves indicate qualitatively the effect of certain elements in imparting creep strength to steels.

14. The Metal-Gas Company reported that, in rupture tests of chromized low-carbon steel, the coating stretched out and continued to give protection against oxidation. Intergranular fissuring still occurred, but the break in the stress-rupture-time curve for tests on unprotected samples was eliminated.

15. At the research establishment of the Department of Aeronautical and Engineering Research a program of research is being conducted on the fundamentals of fracture of metals at prolonged time periods under both static and dynamic loads.

16. The United Steel Companies are studying the effect of notches on the rupture properties of 0.5Mo and Red Fox 33 steels. Substantial reductions in strength were observed (from 20 to 100 percent of the life of unnotched 0.5Mo steel at 1050° F, depending on steel-making practice; 10-to-1 reduction in life of Red Fox 33 at 1112° and 1200° F, the decrease being greater the higher the solution temperature). The reduction in life was not restricted to materials with low elongation in the unnotched condition.

17. At the Mond Nickel Company fundamental programs are in progress on the mechanism of creep. This research has included an extensive study of the movements of grain boundaries and associated grains in materials which creep at room temperature. The work is to be extended to nickel and Nimonic alloys. They subscribe to the creep theory involving grain-boundary viscosity, stress concentration at intersections of grain boundaries, and relaxation of stress concentration.

They believe viscosity effects can be very large. This is suggested as the cause of pronounced effects of deoxidants and other small additions which can effect creep to such a pronounced degree. They suggest that viscosity effects may control onset of third-stage creep and have not been able to detect flaws at the onset of third-stage creep.

Aluminum and titanium have been found to show surprising increases in creep resistance of Nimonic alloys at temperatures above the solubility limit.

18. The relationship between the creep curve and fatigue resistance has been studied by The Mond Nickel Company Development and Research Department Laboratory. They found that fatigue strength falls rapidly after the onset of third-stage creep. Creep up to the start of third-stage creep does not affect fatigue strength.

19. Creep of "compeletely" brittle materials is being studied by the Royal Aircraft Establishment. Single and multicrystalline alumina (Al_2O_3) is being used. The work is to be coordinated with self-diffusion studies to determine if there is a relationship between the two. The diffusion work is also a start on the fundamentals of sintering.

20. X-ray investigation of the creep process is in progress at the Royal Aircraft Establishment under Dr. Greenough. The observations of

Dr. W. A. Wood in Australia on aluminum have been checked, but interpreted using the dislocation theory. A nearly perfect solid solution of silver in aluminum did not change polygonization. Splitting of Laue diffraction spots during creep of aluminum containing precipitates could not be detected, indicating that polygonization does not occur in the presence of precipitates.

The work is being extended to fatigue, with preliminary results indicating that polygonites formed by fatigue have much greater (10° to 30°) differences in orientation than those formed by creep.

University of Michigan
Ann Arbor, Mich.

and

Battelle Memorial Institute
Columbus, Ohio
March 1, 1951

APPENDIX

DATA SHEETS ON GAS-TURBINE BLADE MATERIALS

CHEMICAL COMPOSITION OF MATERIALS

Alloy	Chemical composition (percent)												
	C	Mn	Si	Cr	Ni	Co	Mo	W	Cb	V	Ti	Al	Fe
Nimonic 90	0.1 max.	1.0 max.	1.5 max.	18-21	Bal.	15-21	----	---	----	----	1.8-2.7 (Aim 2.4)	0.8-1.8 (Aim 1.2)	5.0 max. (Usual 0.3)
Nimonic 80A	.1 max.	1.0 max.	1.0 max.	18-21	Bal.	2.0 max.	----	---	----	----	1.8-2.7	0.5-1.8	5.0 max.
Nimonic 80	.1 max.	1.0 max.	1.0 max.	18-21	Bal.	2.0 max.	----	---	----	----	1.8-2.7	0.5-1.8	5.0 max.
G18B	.4	.8	1.0	13	13	10	2	2.5	2.0	----	--------	--------	Bal.
R.ex 326	.23	2.93	.74	14.68	19.65	6.80	2.69	---	2.08	----	--------	--------	Bal.
FCB(T)	.11	.41	.50	17.84	12.0	----------	----	---	1.0	----	--------	--------	Bal.
Mo-V	.19	.53	.36	.32	.22	----------	.72	---	----	0.31	--------	--------	Bal.
G32	.3	.8	.3	19	11	Bal.	2	---	1.2	2.8	--------	--------	15

NACA

NIMONIC 90

Chemical composition, percent:

C, 0.10; Mn, 1.0 max.; Si, 1.5 max.; Cr, 18-21; Co, 15-21; Ti, 1.8-2.7 (aim 2.4); Al, 0.8-1.8 (aim 1.2); Fe, 5.0 max.; Ni bal.

Heat treatment:

Solution-treated at 1975° F for 8 hr, air-cooled; aged at 1290° F for 16 hr, air-cooled

TYPICAL MECHANICAL PROPERTIES AT ELEVATED TEMPERATURES

Property	Stress (psi)			
	1200° F (650° C)	1290° F (700° C)	1380° F (750° C)	1500° F (816° C)
Short-time tensile strength	----------	116.5×10^3	103.0×10^3	76.1×10^3
0.1-percent-offset yield strength	----------	83.0	71.7	58.2
Young's Modulus	----------	25,000.0	24,000.0	21,000.0
0.1 percent creep strain in 100 hr	69.5×10^3	53.8	38.0	22.4
0.2 percent creep strain in 100 hr	74.0	58.2	41.5	24.6
0.5 percent creep strain in 100 hr	75.0	60.5	43.7	26.9
Rupture in 100 hr	76.1	61.6	44.8	28.0
0.1 percent creep strain in 300 hr	65.0	48.1	32.5	17.9
0.2 percent creep strain in 300 hr	68.2	52.7	35.8	20.2
0.5 percent creep strain in 300 hr	69.5	53.8	38.1	22.4
Rupture in 300 hr	70.6	54.9	39.2	23.5
0.1 percent creep strain in 1000 hr	58.2	42.5	26.9	13.4
0.2 percent creep strain in 1000 hr	61.6	45.9	30.2	15.7
0.5 percent creep strain in 1000 hr	62.7	47.0	32.5	16.8
Rupture in 1000 hr	63.9	48.2	33.6	17.9
0.1 percent creep strain in 5000 hr	50.4	35.8	20.2	6.7
0.2 percent creep strain in 5000 hr	53.7	38.0	22.4	9.0
0.5 percent creep strain in 5000 hr	54.9	39.2	23.5	10.1
Rupture in 5000 hr	56.0	40.3	24.6	11.2

The data given in the foregoing table were obtained by The Mond Nickel Company. The data on Nimonic 90 from the National Physical Laboratory are not yet available in published form, but were made available for inspection. Some limited data were obtained for tests of 1000 hours' duration at temperatures of 1290°, 1380°, and 1500° F (700°, 750°, and 815° C) which indicate that The Mond Nickel Company data for Nimonic are consistently on the conservative side. These comparisons are shown below:

NIMONIC 90

Test temperature	Laboratory	Stress (psi) for -	
		0.2-percent deformation in 1000 hr	Rupture in 1000 hr
1290° F (700° C)	Mond	45.9×10^3	48.2×10^3
	NPL	47.5	49.3
1380° F (750° C)	Mond	30.2	33.6
	NPL	33.6	37.2
1500° F (815° C)	Mond	15.7	17.9
	NPL	17.5	20.2

The following pamphlet on Nimonic 90 was obtained from The Mond Nickel Company.

NIMONIC 90

The information given in this leaflet supplements that published in "The Nimonic Series of Alloys—Their Application to Gas Turbine Design", by making available data on the physical and mechanical properties of the latest addition to the Nimonic Series of Alloys, known as Nimonic 90.

THE DEVELOPMENT of Nimonic 90 is the result of the continuous research in our laboratories to improve the Nimonic Series of Alloys combined with the long experience in the production of this type of alloy of Messrs. Henry Wiggin & Co. Ltd. It makes available to designers a wrought alloy which may be operated at higher temperatures and stresses than the previous best wrought material available, Nimonic 80A.

The increased creep strength available with Nimonic 90 is well shown by a comparison of the creep criteria specified in the appropriate Ministry of Supply Specifications where for the same criterion at the same temperature of 750° C. Nimonic 80A is required to withstand a stress of 17 tons per square inch and Nimonic 90 a stress of 19 tons per square inch. At 815° C. and even 870° C. Nimonic 90 has shown high load-carrying capacity for long periods. The alloy has been well proved by its successful use in the latest types of turbo-jet engines for a considerable period.

The forging of Nimonic 90 can be satisfactorily performed in the range 1050—1150° C. but the alloy rapidly increases in stiffness as the temperature falls below 1000° C. and frequent reheatings are necessary. As for Nimonic 80 and 80A the furnace atmospheres should be oxidizing and the fuel used should have a low sulphur content.

The heat-treatment of Nimonic 90 consists of a solution treatment of 8 hours at 1080° C. followed by air-cooling and reheating to 700° C. for 16 hours and air-cooling.

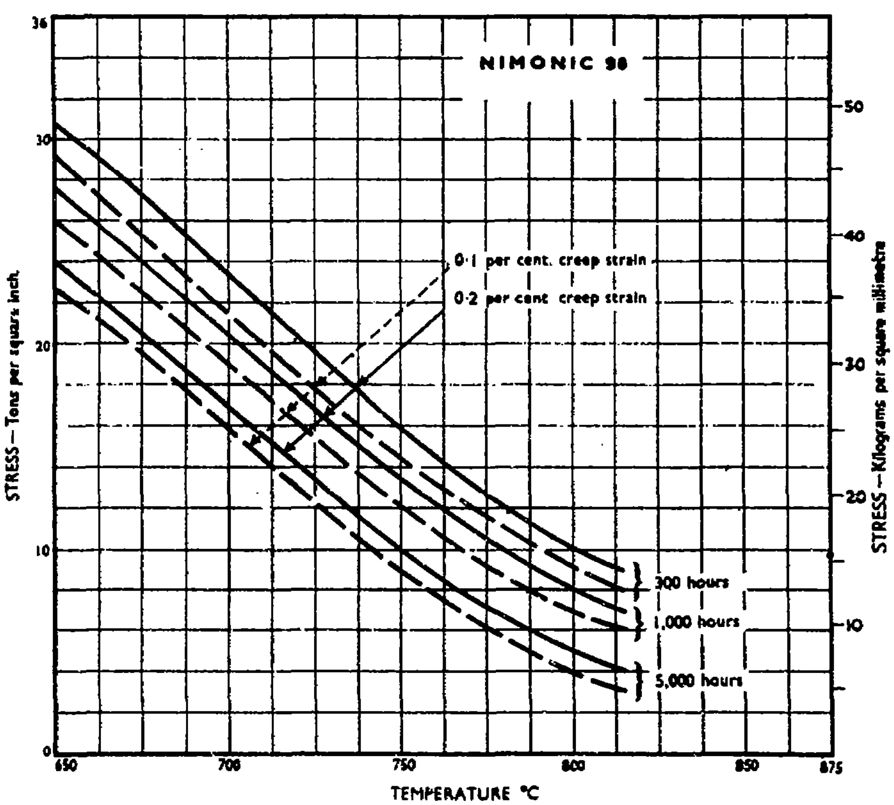
As in the case of Nimonic 80 and 80A, the altered superficial layer of approximately ·003 in. should be removed by machining or other means prior to putting the component into service. As with Nimonic 80 and 80A, the creep-resisting properties of the alloys are adversely affected by cold-work applied subsequent to the solution heat-treatment.

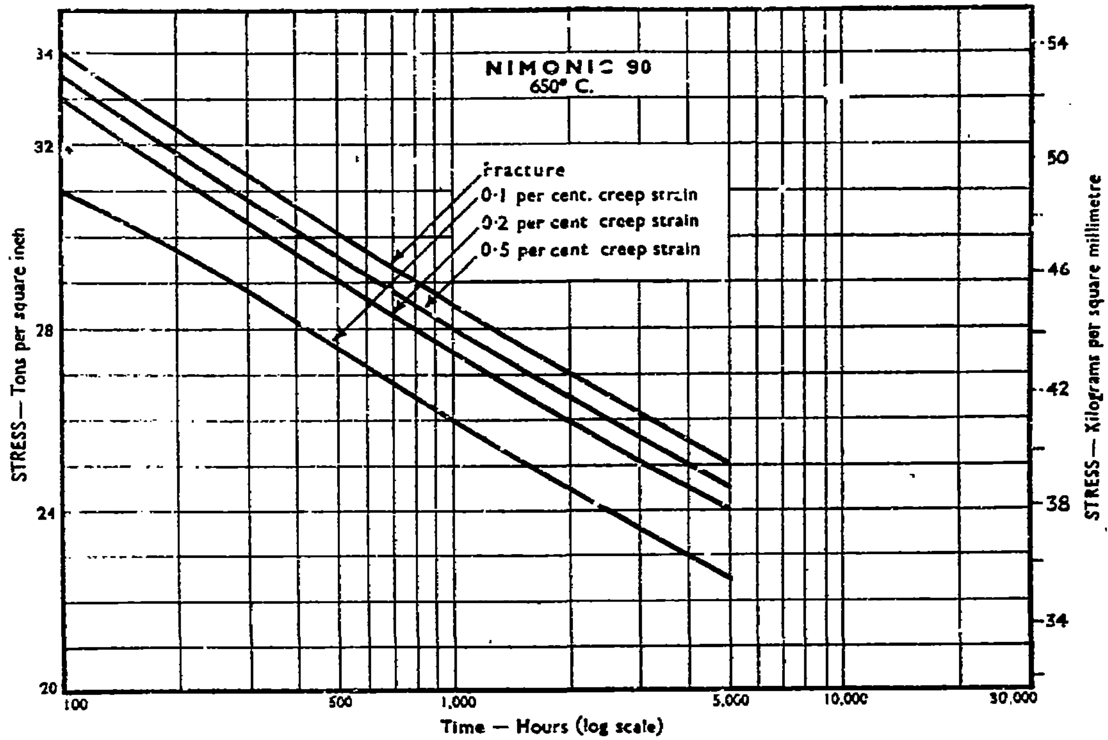
Nimonic 90 is available in bar, sections, forgings, sheet, strip and wire.

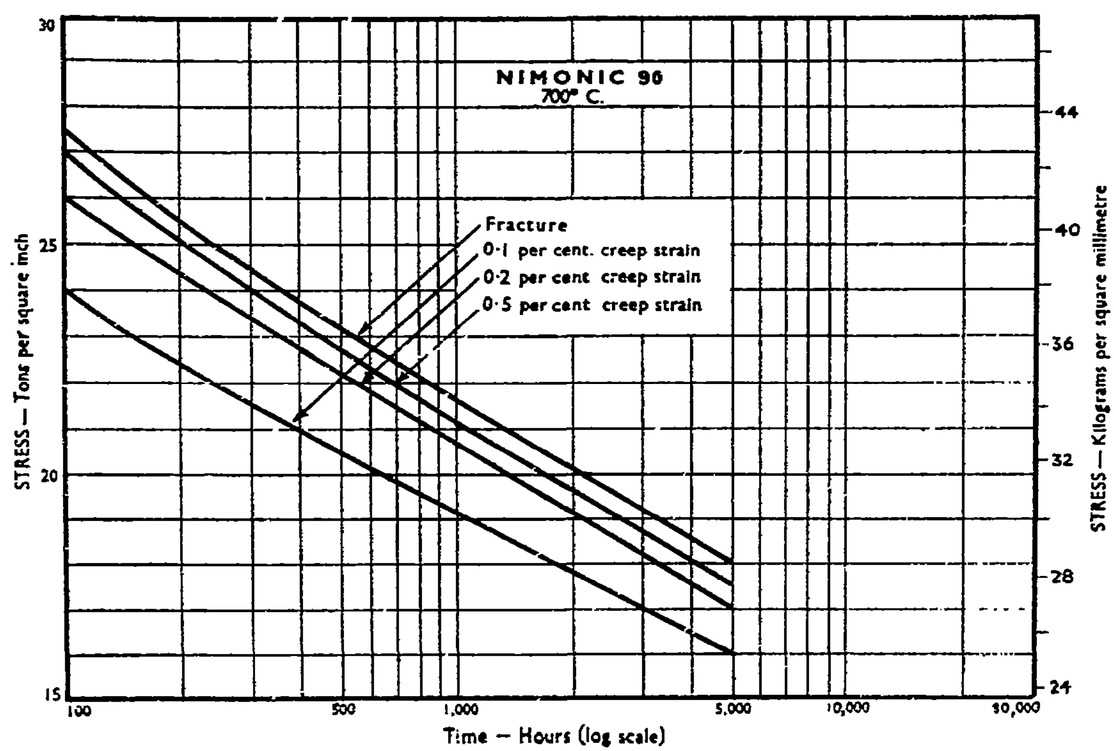
[redacted]

Physical Properties of Nimonic 90

Specific gravity	.. ..	8·27
Mean coefficient of thermal expansion in millionths per °C.:	20.100° C.	11·6
	20–200° C.	12·6
	20–300° C.	12·7
	20–400° C.	13·5
	20–500° C.	13·7
	20–600° C.	14·2
	20–700° C.	15·0
	20–800° C.	16·0
	20–900° C.	17·0
Electrical resistivity	..	Fully heat-treated condition about 115 microhm-cms/cm².




NIMONIC 90
650° C.
Fracture
0·1 per cent. creep strain
0·2 per cent creep strain
0·5 per cent creep strain
STRESS — Tons per square inch
STRESS — Kilograms per square millimetre
Time — Hours (log scale)
20
24
28
32
34
34
38
42
46
50
54
100
500
1,000
5,000
10,000
30,000


NIMONIC 90
700° C.
Fracture
0·1 per cent. creep strain
0·2 per cent creep strain
0·5 per cent creep strain
STRESS — Tons per square inch
STRESS — Kilograms per square millimetre
Time — Hours (log scale)
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20
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30
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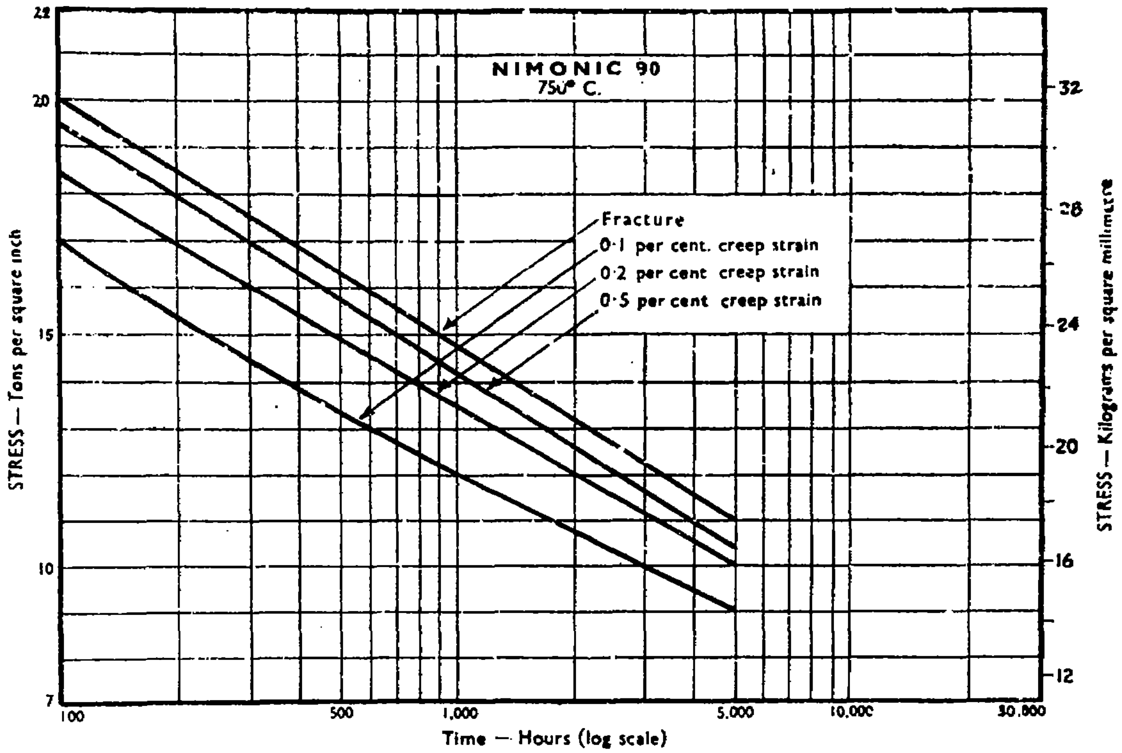
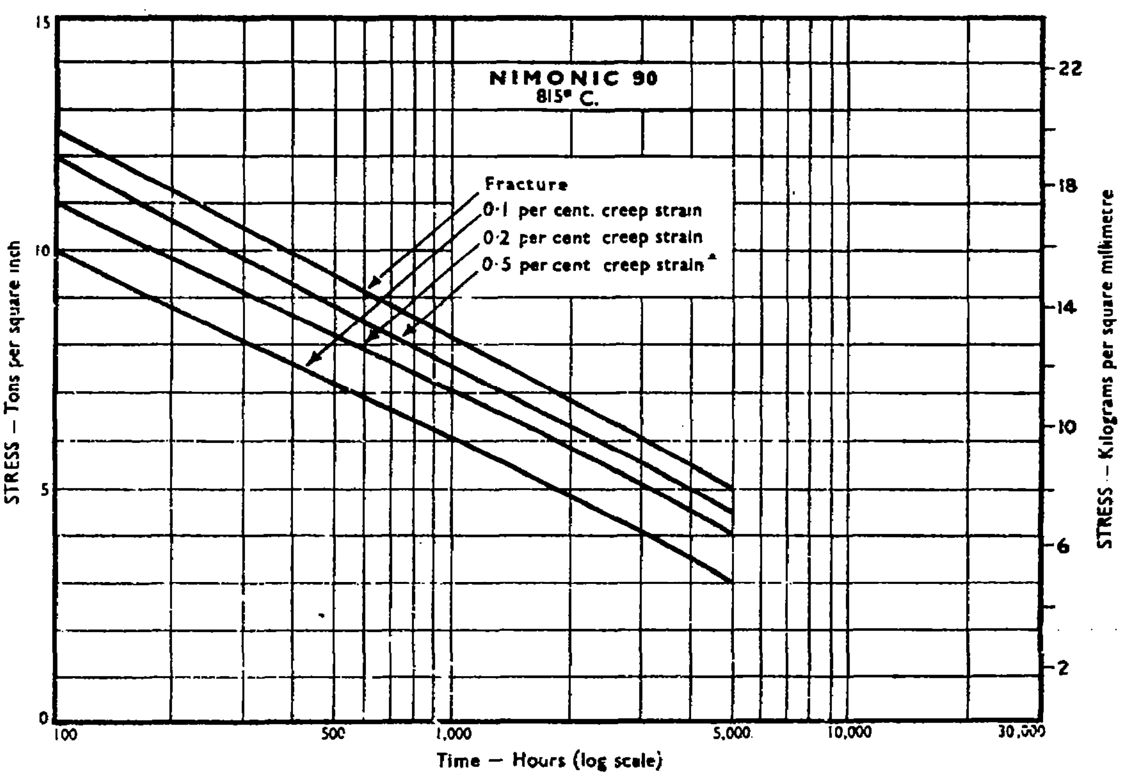
40
44
100
500
1,000
5,000
10,000
30,000


NIMONIC 90
750° C.
Fracture
0·1 per cent. creep strain
0·2 per cent creep strain
0·5 per cent creep strain
STRESS — Tons per square inch
STRESS — Kilograms per square millimetre
Time — Hours (log scale)


NIMONIC 90
815° C.
Fracture
0·1 per cent. creep strain
0·2 per cent creep strain
0·5 per cent creep strain
STRESS — Tons per square inch
STRESS — Kilograms per square millimetre
Time — Hours (log scale)

[redacted]

NIMONIC 80A

Chemical composition, percent:

C, 0.10 max.; Mn, 1.0 max.; Si, 1.0 max.; Cr, 18-21; Co, 2.0 max.; Ti, 1.8-2.7; Al, 0.5-1.8; Fe, 5.0 max.; Ni, bal.

Heat treatment:

Solution-treated at 1975° F for 8 hr, air-cooled; aged at 1290° F for 16 hr, air-cooled

TYPICAL MECHANICAL PROPERTIES AT ELEVATED TEMPERATURES

Property	Stress (psi)		
	1200° F (650° C)	1290° F (700° C)	1380° F (750° C)
0.1 percent creep strain in 100 hr	58.2×10^3	42.6×10^3	32.5×10^3
0.2 percent creep strain in 100 hr	63.8	51.5	35.8
0.5 percent creep strain in 100 hr	66.1	54.9	38.1
Rupture in 100 hr	67.2	57.1	39.2
0.1 percent creep strain in 300 hr	51.5	34.7	25.8
0.2 percent creep strain in 300 hr	58.2	43.7	30.2
0.5 percent creep strain in 300 hr	60.5	48.2	32.5
Rupture in 300 hr	61.6	49.3	33.6
0.1 percent creep strain in 1000 hr	44.8	25.8	17.9
0.2 percent creep strain in 1000 hr	50.4	34.7	23.5
0.5 percent creep strain in 1000 hr	53.8	39.2	24.6
Rupture in 1000 hr	54.9	40.3	25.8
0.1 percent creep strain in 5000 hr	35.8	13.4	10.1
0.2 percent creep strain in 5000 hr	41.4	22.4	14.5
0.5 percent creep strain in 5000 hr	44.8	26.9	15.7
Rupture in 5000 hr	45.9	28.0	16.8
0.1 percent creep strain in 10,000 hr	----	----	----
0.2 percent creep strain in 10,000 hr	----	17.9	10.5
0.5 percent creep strain in 10,000 hr	40.4	21.3	12.3
Rupture in 10,000 hr	41.4	23.5	(13.4)

NIMONIC 80

Chemical composition, percent:

C, 0.10 max.; Mn, 1.0 max.; Si, 1.0 max.; Cr, 18-21; Co, 2.0 max.; Ti, 1.8-2.7; Al, 0.5-1.8; Fe, 5.0 max.; Ni, bal.

Heat treatment:

Solution-treated at 1975° F for 8 hr, air-cooled; aged at 1290° F for 16 hr, air-cooled

TYPICAL MECHANICAL PROPERTIES AT ELEVATED TEMPERATURES

Property	Stress (psi)			
	1110° F (600° C)	1200° F (650° C)	1290° F (700° C)	1380° F (750° C)
Short-time tensile stength	----------	----------	85.2×10^3	----------
0.1-percent-offset yield strength	----------	----------	78.4	----------
Young's modulus	----------	----------	54,000.0	----------
0.1 percent creep strain in 100 hr	66.1×10^3	45.9×10^3	37.0	22.4×10^3
0.2 percent creep strain in 100 hr	71.7	54.9	42.6	28.0
0.5 percent creep strain in 100 hr	73.9	59.4	45.9	32.5
Rupture in 100 hr	75.0	62.7	49.3	33.6
0.1 percent creep strain in 300 hr	57.1	42.6	31.4	17.9
0.2 percent creep strain in 300 hr	65.0	49.3	37.0	21.3
0.5 percent creep strain in 300 hr	69.4	54.9	40.3	25.8
Rupture in 300 hr	72.8	57.1	42.6	29.1
0.1 percent creep strain in 1000 hr	49.3	37.0	24.6	13.4
0.2 percent creep strain in 1000 hr	56.0	43.7	30.2	15.7
0.5 percent creep strain in 1000 hr	62.7	48.1	33.6	17.9
Rupture in 1000 hr	66.1	50.4	35.8	21.3
0.1 percent creep strain in 5000 hr	40.3	30.2	----------	----------
0.2 percent creep strain in 5000 hr	47.0	34.7	20.2	(9)
0.5 percent creep strain in 5000 hr	52.6	39.2	23.5	11.2
Rupture in 5000 hr	54.9	40.3	25.7	12.3
0.1 percent creep strain in 10,000 hr	38.1	26.9	----------	----------
0.2 percent creep strain in 10,000 hr	43.7	31.4	16.8	7.8
0.5 percent creep strain in 10,000 hr	49.3	34.7	20.2	9.0
Rupture in 10,000 hr	51.5	35.8	21.3	10.1

NACA

G18B

Heat treatment:

Solution-treated at 2370° F

TYPICAL MECHANICAL PROPERTIES AT ELEVATED TEMPERATURES

Property	Stress (psi) (1)							
	1110° F (600° C)	1200° F (650° C)	1290° F (700° C)	1380° F (750° C)	1470° F (800° C)	1560° F (850° C)	1650° F (900° C)	1740° F (950° C)
Short-time tensile strength	89.5×10^3	69.5×10^3	62.7×10^3	58.3×10^3	49.3×10^3	40.3×10^3	31.4×10^3	----------
0.1 percent strain in 300 hr	----------	17.9	11	8.1	----------	----------	----------	----------
0.5 percent strain in 300 hr	----------	29.1	20.8	14.3	11.9	----------	----------	----------
1.0 percent strain in 300 hr	----------	33.6	25.1	17.7	12.8	----------	----------	----------
Rupture in 300 hr	----------	41.6	30.9	18.4	14.3	10.3	7.8	3.9×10^3
0.1 percent strain in 1000 hr	----------	14.8	8.1	7.4	----------	----------	----------	----------
0.5 percent strain in 1000 hr	----------	25.6	17.5	13.2	----------	----------	----------	----------
1.0 percent strain in 1000 hr	----------	30.5	21.	14.8	11.2	----------	----------	----------
Rupture in 1000 hr	----------	35.8	25.6	16.4	12.5	9.0	7.2	3.4
0.1 percent strain in 3000 hr	----------	13.0	6.0	6.7	----------	----------	----------	----------
0.5 percent strain in 3000 hr	----------	17.5	15.2	12.1	----------	----------	----------	----------
1.0 percent strain in 3000 hr	----------	25.6	18.4	12.5	----------	----------	----------	----------
Rupture in 3000 hr	----------	30.9	22.2	14.6	----------	----------	----------	----------
0.1 percent strain in 10,000 hr	----------	(11.2)	----------	5.6	----------	----------	----------	----------
0.5 percent strain in 10,000 hr	----------	14.3	13.4	10.1	----------	----------	----------	----------
1.0 percent strain in 10,000 hr	----------	(20.2)	15.7	12.3	9.9	----------	----------	----------
Rupture in 10,000 hr	----------	26.4	18.8	14.0	10.1	6.7	5.0	2.8
0.1 percent strain in 30,000 hr	----------	----------	----------	----------	----------	----------	----------	----------
0.5 percent strain in 30,000 hr	----------	(13.0)	(12.1)	----------	----------	----------	----------	----------
1.0 percent strain in 30,000 hr	----------	(15.9)	(13.4)	----------	----------	----------	----------	----------
Rupture in 30,000 hr	----------	(22.4)	16.1	----------	----------	----------	----------	----------

[1]All figures quoted are typical average values, but are expected to be reproducible within about 10 percent.

R.ex 326

Chemical composition, percent:

C, 0.25; Mn, 3.0; Cr, 17.0; Ni, 17.0; Co, 7.0; Mo, 2.5; Cb, 1.8

Heat treatment:

Solution-treated at 2280° F, water-quenched and aged

Property	Stress (psi)	
	1200° F (650° C) (1)	1290° F (700° C)
Short-time tensile strength	57.4×10^3	51.6×10^3
0.1-percent-offset yield strength	36.8	35.2
0.1 percent plastic strain in 300 hr	22.4	15.9
0.2 percent plastic strain in 300 hr	26.9	19.7
0.5 percent plastic strain in 300 hr	30.2	25.6
Rupture in 300 hr	35.8-42.5	24.6-29.1
0.1 percent plastic strain in 1000 hr	20.2	13.4
0.2 percent plastic strain in 1000 hr	24.6	15.9
0.5 percent plastic strain in 1000 hr	28.0	21.2
Rupture in 1000 hr	31.4-38.0	22.4-25.8
0.1 percent plastic strain in 5000 hr	14.6	----------------
0.2 percent plastic strain in 5000 hr	17.9	10.1
0.5 percent plastic strain in 5000 hr	22.4	14.6
Rupture in 5000 hr	25.8-29.1	----------------
0.1 percent plastic strain in 10,000 hr	12.3	9.6
0.2 percent plastic strain in 10,000 hr	15.7	11.0
0.5 percent plastic strain in 10,000 hr	19.0	13.2
Rupture in 10,000 hr	>21.2	15.7

[1]The data at 1200° F are compiled, in the main, from transverse tests taken from the rim of a shaft forging. The data for 1290° F were obtained on bar stock.

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FCB(T)

Chemical composition, percent:

C, 0.12; Cr, 17.5; Ni, 12.00; Cb, 1.0

Heat treatment:

Hot-rolled or forged

TYPICAL MECHANICAL PROPERTIES AT ELEVATED TEMPERATURES

[Data for bar stock]

Property	Stress (psi)	
	1110° F (600° C)	1200° F (650° C)
Short-time tensile strength	55.0×10^3	53.5×10^3
0.1-percent-offset yield strength[1]	20.3	22.3
0.1 percent plastic strain in 300 hr	26.9	15.7
0.2 percent plastic strain in 300 hr	29.1	17.9
0.5 percent plastic strain in 300 hr	33.6+	20.2
Rupture in 300 hr	40.3-42.5	22.4-24.6
0.1 percent plastic strain in 1000 hr	21.3	11.2
0.2 percent plastic strain in 1000 hr	23.5	14.6
0.5 percent plastic strain in 1000 hr	26.9	16.8
Rupture in 1000 hr	31.4-33.6	17.9-20.2
0.1 percent plastic strain in 5000 hr	17.9	6.7
0.2 percent plastic strain in 5000 hr	17.9+	6.7
0.5 percent plastic strain in 5000 hr	20.2	9.0
Rupture in 5000 hr	22.4-24.6	11.2-13.4
0.1 percent plastic strain in 10,000	15.7	4.5
0.2 percent plastic strain in 10,000	16.8	5.6
0.5 percent plastic strain in 10,000	17.9	6.7
Rupture in 10,000 hr	>20.2	9.0

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[1]A safe working stress at 1020° F appears to be at about the 0.1-percent-offset yield strength of about 20,000 psi.

JESSOP G32 STEEL

Chemical composition, percent:

C, 0.3; Mn, 0.8; Si, 0.3; Cr, 19; Ni, 12; Co, 45; Mo, 2.0; Cb, 1.2; V, 2.8; Fe, 16 approx.

Heat treatment:

Oil-quenched from 2335° F (1280° C) after 10-min soak; aged for 46 hr at 1380° F (750° C)

TYPICAL MECHANICAL PROPERTIES AT ELEVATED TEMPERATURES

Property	Stress (psi)			
	1290° F (700° C)	1380° F (750° C)	1470° F (800° C)	1560° F (850° C)
Rupture in 30 hr	53.8×10^3	47.0×10^3	35.8×10^3	26.9×10^3
Rupture in 100 hr	49.3	40.3	29.1	22.4
Rupture in 300 hr	45.9	37.0	26.9	20.2
Rupture in 1000 hr	41.5	33.6	22.4	15.7
Stress for 0.5 percent strain in 300 hr	----------	32.5	17.9	13.4
Stress for 1.0 percent strain in 300 hr	----------	33.6	22.4	17.9

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JESSOP R20 STEEL

Chemical composition, percent:

C, 0.15; Mn, 0.80; Si, 0.30; Cr, 19.0; Ni, 14.0; Cb, 1.7

Heat treatment:

This material is heat-treated by cooling in air from 2010° F (1100° C); no marked variation of creep strength is found between small bar samples and large forgings

TYPICAL MECHANICAL PROPERTIES AT ELEVATED TEMPERATURES

Property	Stress (psi)			
	1020° F (550° C)	1110° F (600° C)	1200° F (650° C)	1290° F (700° C)
Rupture in 100 hr	$(47) \times 10^3$	42.5×10^3	31.4×10^3	20.2×10^3
Rupture in 300 hr	(42.5)	39.2	25.8	15.7
Rupture in 1000 hr	40.3	31.4	19.1	12.3
Rupture in 3000 hr	37.0	23.5	13 4	9.0
Rupture in 10,000 hr	(24.6)	19.1	10.1	----------

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R.ex 448

Chemical composition, percent:

Cr, 12.0; Mo, 0.5; V, 0.25; Cb, 0.4

Heat treatment:

Hardenable, details not available

Room-temperature mechanical properties - bar stock:

0.05 percent proof stress, psi 114,000
0.10 percent proof stress, psi 123,000
0.20 percent proof stress, psi -------
0.50 percent proof stress, psi 137,000
Tensile strength, psi 157,000
Elongation, percent 18
Reduction in area, percent 46

TYPICAL MECHANICAL PROPERTIES AT ELEVATED TEMPERATURES

Property[1]	Stress (psi)		
	1020° F (550° C)	1110° F (600° C)	1200° F (650° C)
0.2 percent creep strain in 300 hr	----------	31.4×10^3	19.0×10^3
0.5 percent creep strain in 300 hr	----------	38.1	23.5
Rupture in 300 hr	62.8×10^3	40.3	30.2
0.2 percent creep strain in 1000 hr	----------	22.4	15.7
0.5 percent creep strain in 1000 hr	----------	33.6	17.9
Rupture in 1000 hr	58.3	38.1	>22.4

[1]Rupture properties on bar stock - no 10,000-hr data are yet available, but a test at 1110° F (600° C) and 15,700 psi has completed 3217 hr with a creep deformation of 0.161 percent.

[redacted]

JESSOP H46 STEEL

(Developed primarily as a disk material)

Chemical composition, percent:

C, 0.2; Mn, 0.4; Si, 0.3; Cr, 11.0; Mo, 0.5; V, 0.7; Cb, 0.15

Heat treatment:

Hardenable, details not available

Room-temperature mechanical properties:

0.05 percent proof stress, psi 89,500
0.10 percent proof stress, psi 94,000
0.20 percent proof stress, psi 101,000
0.50 percent proof stress, psi 112,000
Tensile strength, psi 135,000
Elongation, percent 17
Reduction of area, percent 45

TYPICAL MECHANICAL PROPERTIES AT ELEVATED TEMPERATURES

Property	Stress (psi) (1)			
	1020° F (550° C)	1110° F (600° C)	1200° F (650° C)	1290° F (700° C)
Rupture in 100 hr	(62.6) × 10³	42.6 × 10³	29.9 × 10³	15.6 × 10³
0.5 percent strain in 300 hr	------------	27.0	14.2	----------
1.0 percent strain in 300 hr	------------	31.3	22.8	6.4
Rupture in 300 hr	58.3	28.4	27.8	11.4
0.5 percent strain in 1000 hr	------------	22.0	12.8	----------
1.0 percent strain in 1000 hr	------------	27.0	15.6	----------
Rupture in 1000 hr	(54)	31.3	19.9	7.8
0.5 percent strain in 3000 hr	------------	19.2	----------	----------
Rupture in 3000 hr	------------	25.6	15.6	(5.7)
0.5 percent strain in 10,000 hr	------------	17.1	----------	----------
Minimum creep rate, 10^{-5} in./in./hr	51.2	27.0	17.8	5.7
Minimum creep rate, 10^{-6} in./in./hr	39.8	22.0	10.0	----------

[1]These data were obtained on bar material in the hardened and tempered condition. Creep properties obtainable on turbine disk forgings are usually about 10 to 20 percent lower than these figures, depending on the size of the forgings and the creep criterions considered. All figures quoted are typical values and should not be regarded as specification.

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TABLE I.- DISK ALLOYS FOR GAS TURBINES

Alloy	Chemical composition (percent)									
	C	Mn	Si	Cr	Ni	Co	Mo	W	Cb	V
Mo-V	0.15	0.4	0.15	-----	----	----	0.5	---	----	0.25
1Cr-0.75Mo	.35	---	----	1.25	----	----	.75	---	----	----
3Cr-Mo-W-V	.2	.3	.4	3.0	----	----	.6	0.5	----	.8
H46	.2	.4	.3	11.0	----	----	.5	---	0.15	.7
[a]"Improved" H46	----	---	----	-----	----	----	----	---	----	----
[a]R.ex 448	----	---	----	-----	----	----	----	---	----	----
18-12+Cb	.12	.4	.5	17.5	12.0	----	----	---	1.0	----
G18B	.4	.8	1.0	13.0	13.0	10.0	2.0	2.5	3.0	----
R.ex 326	.25	3.0	.8	17.0	17.0	7.0	2.5	---	1.8	----

[a]Composition not disclosed.

TABLE II.- HEAT TREATMENTS FOR DISK ALLOYS

Alloy	Quenching treatment		Tempering treatment		Comments
	Temperature (°F)	Cooling (1)	Temperature (°F)	Time (hr)	
Mo-V	1740	O.Q.	1275	4	Treatment for aircraft turbine disks
1Cr-0.75Mo	1580	O.Q.	1166	---	Treatment for aircraft turbine disks
3Cr-Mo-W-V	1940	O.Q.	1292	---	Usually air-cooled from 2100° F before hardening; actual quenching temperatures vary from 1900° to 1940° F and tempering temperatures from 1256° to 1328° F, depending on the manufacturer and the application requirements
H46	2100	O.Q.	1295	---	--
R.ex 448					Treatment has not been disclosed by producer
18-12+Cb	2012	A.C.	----	---	Large rotors are also water-quenched
G18B (A)	2370	----	----	---	--
(B)	2370	----	1472	---	--
(C)	2370	----	(2)	(2)	Usual yield strength of 56,000 psi for 0.1-percent offset; in practice, equipment limitations may result in considerably lower values for the larger disks
R.ex 326	2280	W.Q.	----	---	Treatment has been used on large forging for rotors

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[1]O.Q., oil-quenched; W.Q., water-quenched.
[2]Warm-worked at 1380° F.

TABLE III.- ROOM-TEMPERATURE PHYSICAL PROPERTIES OF DISK ALLOYS

Alloy	Tensile strength (psi)	Offset yield strengths (psi)			Elongation (percent)	Reduction of area (percent)	Type of material tested
		0.1 percent	0.2 percent	0.5 percent			
Mo-V	130,000	116,000	116,000	-------	23	68	Bar stock (0.2c); oil-quenched 1742° F; tempered 1202° F
	133,500	-------	-------	115,000	19	61	25-in.-diam. by 2-in.-thick disk; tangential at hub
1Cr-0.75Mo	130,000	96,000	-------	-------	20	40	Oil-quenched 1580° F; tempered 1166° F
	150,000	132,000	134,000	134,000	22	64	---------- ? ----------
3Cr-Mo-W-V	127,500	107,500	-------	115,000	23	69	Rim } 25-in.-diam. by 2-in.-thick disk; oil-quenched 1900° F; tempered 1274° F
	123,000	103,000	-------	108,000	16	54	Hub
	124,000	112,000	-------	112,000	20	59.5	Center
	123,000	100,000	103,000	108,000	20	55	Oil-quenched 1922° F; tempered 1328° F
	139,000	115,000	118,000	121,000	19	53	Oil-quenched 1922° F; tempered 1256° F
	144,000	122,000	126,000	130,000	21	60	Air-cooled 2102° F; oil-quenched 1940° F; tempered 1292° F
12Cr (H46)	135,000	94,000	105,000	-------	17	45	Turbine disk forging
12Cr (R.ex 448)	157,000	122,900	-------	137,000	18	46	$\frac{3}{4}$- in. bar stock
	141,000	117,000	-------	129,000	18.5	51	Rim-tangential } 20-in.-diam. by $3\frac{1}{2}$-in.-thick disk
	136,000	-------	-------	122,500	11.5	21	Center-edge
	139,000	-------	-------	132,000	17	47	Center of section
	138,000	-------	-------	125,000	20	48	Rim-tangential } 23-in.-diam. by $2\frac{3}{4}$-in.-thick disk
	135,500	-------	-------	124,000	15	37	Center of section
18-12+Cb	91,500	29,800	34,000	38,500	58	65	Bar-stock data
	87,500	32,700	35,600	39,200	48	62	Disk data
G18B	103,000	31,400	38,000	44,800	40	45	Solution-treated 2372° F
	107,500	40,300	47,000	56,000	25	22	Solution-treated 2372° F; aged at 1472° F
	107,500	56,000	62,750	69,500	12	15	Solution-treated 2372° F; warm-worked at 1364° F
R.ex 326	95,000	50,000	55,000	58,000	15-40	20-48	Rotor data
	100,000	53,800	58,750	65,000	35	49	Bar stock

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TABLE IV.- COMPARATIVE PROPERTIES OF BRITISH DISK ALLOYS AT HIGH TEMPERATURES

Alloy	Temperature (°F)	Stress (psi) for indicated creep strain				Rupture strength (psi)			Tests on aircraft disks	
		300 hr		1000 hr		100 hr	300 hr	1000 hr	Stress (psi)	Deformation in 300 hr (percent)
		0.2 percent	0.5 percent	0.2 percent	0.5 percent					
Mo-V	1022	26,900	------	22,400	------	------	49,300	38,000	------	------
1Cr-0.75Mo		------	------	------	------	------	------	17,900	35,800	1.0
[a]3Cr-Mo-W-V		21,500	29,800	19,000	27,500	------	------	------	------	------
[b]3Cr-Mo-W-V		40,000	44,000	34,000	42,000	------	64,000	58,000	47,000-59,000	1.0
H46		------	------	------	------	62,500	58,400	53,000	------	------
R.ex 448		------	------	------	------	------	62,700	58,250	------	------
Mo-V	1112	17,900	------	12,300	------	------	28,000	22,400	24,600	0.13
1Cr-0.75Mo		------	------	------	------	------	------	9,000	14,100	1.0
[a]3Cr-Mo-W-V		20,000	26,400	16,750	22,500	------	------	------	------	------
[b]3Cr-Mo-W-V		27,000	35,400	18,000	28,000	------	40,000	34,000	24,600	0.196-0.246
H46		------	27,000	------	22,000	42,750	37,400	31,300	31,400	1.0
"Improved" H46		------	------	------	------	------	------	------	40,000-42,500	1.0
R.ex 448		31,400	38,100	22,400	33,600	------	>40,300	38,000	33,600	0.190-0.219
18-12+Cb		29,000	22,000-34,000	23,500	21,000-27,000	42,500	39,000-42,500	30,000-33,600	------	------
[b]3Cr-Mo-W-V	1202	------	------	------	------	22,400	22,000	15,500	14,000-19,000	1.0
H46		------	14,223	------	12,800	29,900	27,700	19,900	22,850	1.0
R.ex 448		19,000	23,500	15,700	17,900	------	30,250	>22,400	------	------
18-12+Cb		18,000	20,000	14,500	14,000-16,800	31,400	22,000-26,000	18,000-20,000	------	------
[c]G18B		------	29,000	------	25,000	------	41,750	35,800	------	------
[d]G18B		------	------	------	36,000	------	------	------	------	------
R.ex 326		27,000	30,000	24,600	28,000	------	36,000-42,500	31,000-38,600	------	------

[a]Tempered at 1328° F to 112,000-123,000 psi tensile strength.
[b]Tempered at 1256° F to 134,000-146,000 psi tensile strength.
[c]Solution-treated.
[d]Solution-treated and warm-worked disks.

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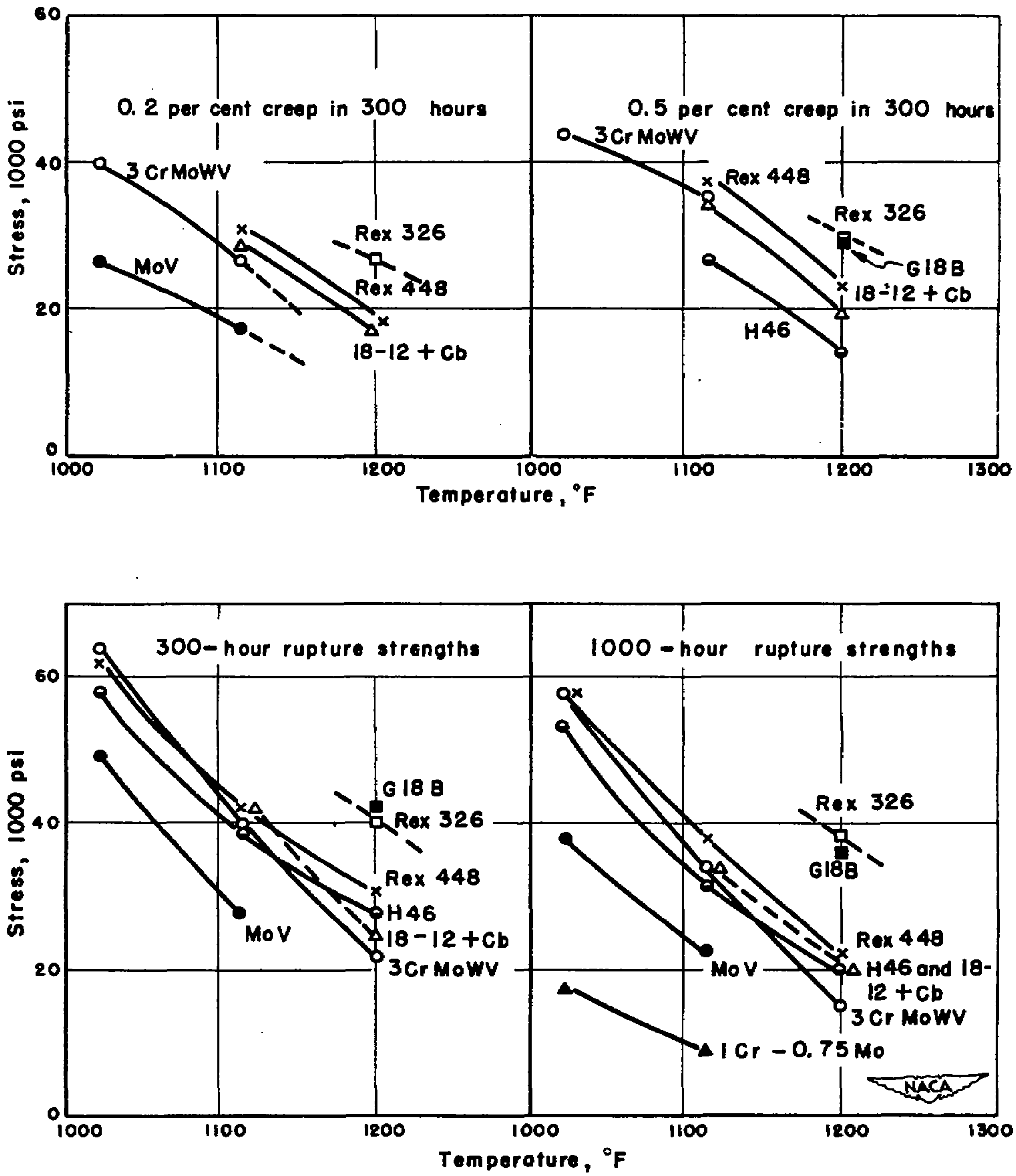

FIGURE 1. COMPARATIVE RUPTURE AND CREEP PROPERTIES FOR DISK ALLOYS